metal treatment

Vol. 26 : No. 168 SEPTEMBER, 1959 Price 2/6

Heat Treatment

in salt bath furnaces, as in greenhouses, brings its rewards. The 'Cassel' Heat Treatment Service offers you the benefits of its experience in carburising, heat treatment, tempering, martempering and austempering.

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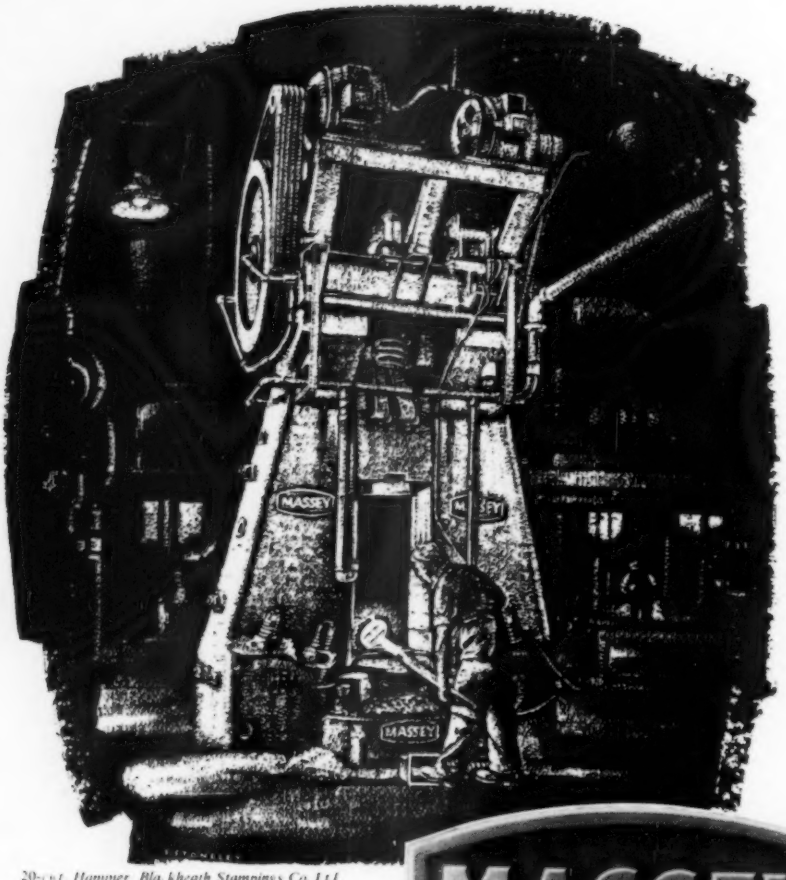

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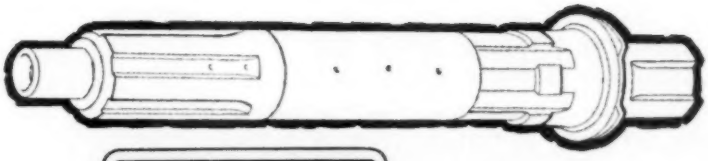

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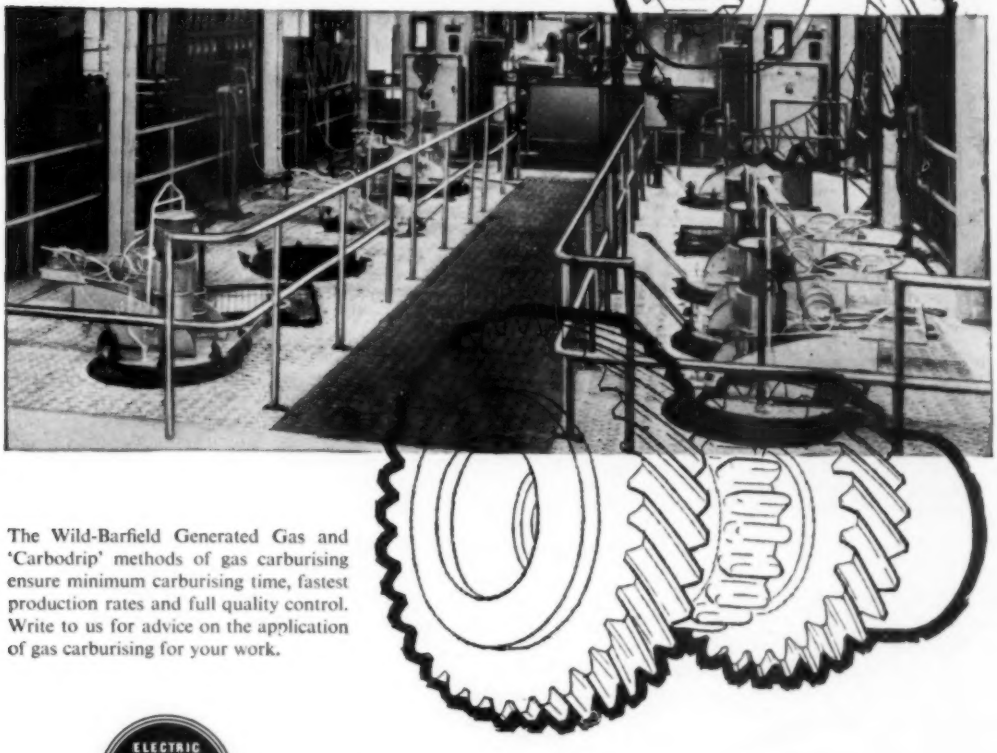

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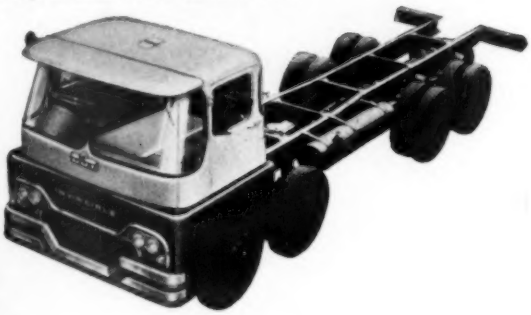
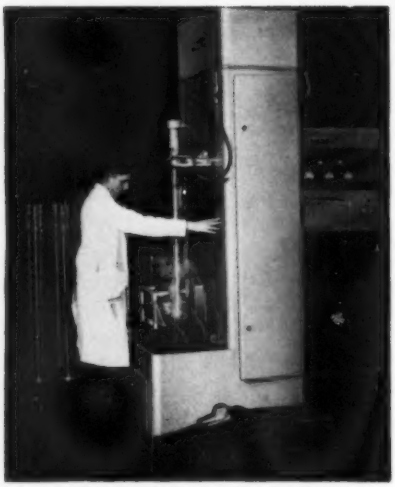
Guy Motors make wide use of Delapena Induction Hardening Equipment

One Delapena-EMA Shaft Hardening Fixture coupled to an E.25 Generator is in non-stop use at Guy Motors, Wolverhampton.





SOME OF THEIR latest heavy vehicle chassis, such as the Invincible Mk. II multi-axled goods chassis shown, incorporate thirty or more parts treated by Delapena induction heating, including:

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POSITION _____

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The electric induction method can solve your melting problem, and G.W.B. can advise you in your choice of the correct induction furnace.

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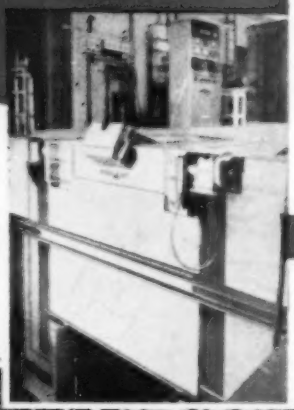

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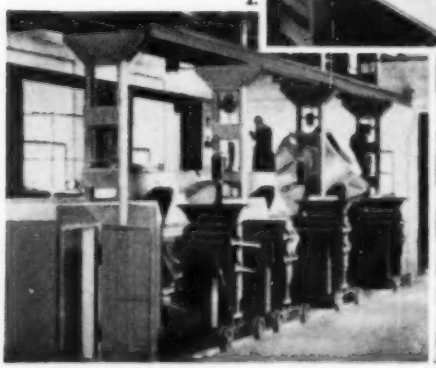

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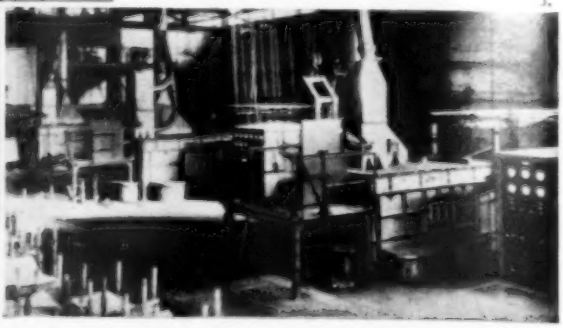

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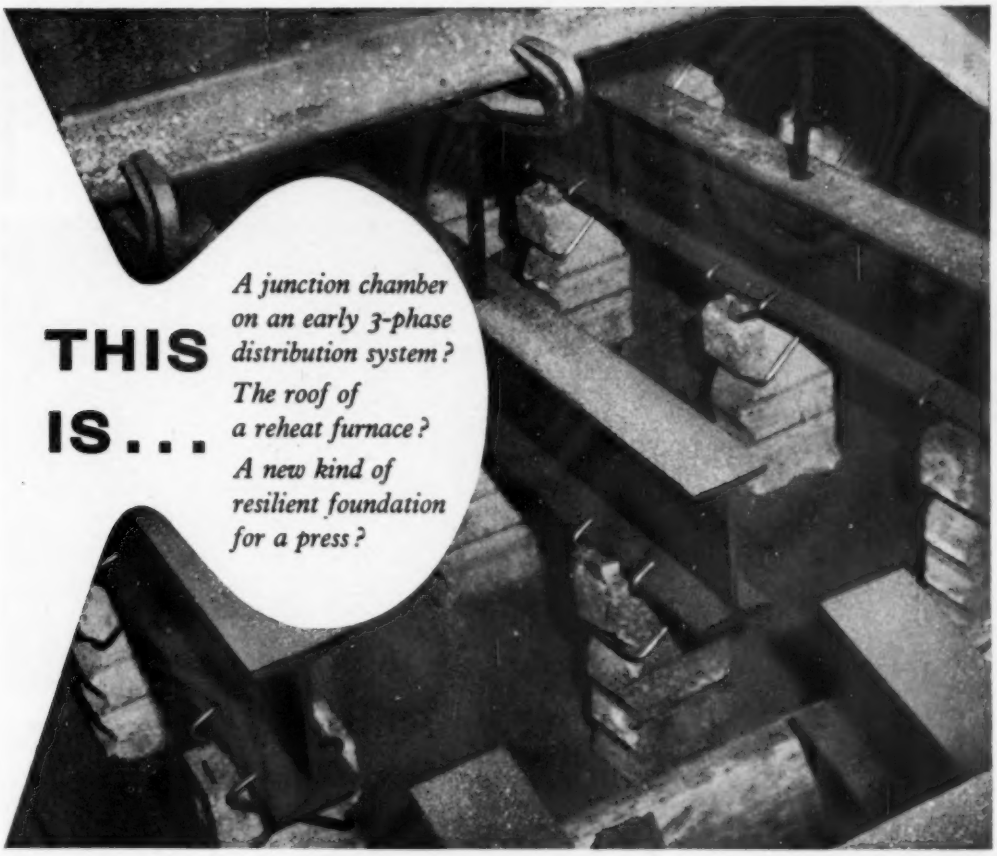
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A junction chamber on an early 3-phase distribution system?
The roof of a reheat furnace?
A new kind of resilient foundation for a press?

Number two is the right answer. It is part of the monolithic roof of a reheat furnace at the Margam works of the Steel Company of Wales, immediately before the refractory was placed. The special anchor bricks hanging from the R.S.J's are secured by heat-resisting clips and will provide support through the full thickness of the refractory lining.
This roof was cast in "TRI-MOR" High Temperature Castable. It is a good example of the type of furnace lining which "TRI-MOR" refractories and Morgan's whole-hearted belief in monolithic structures have made possible.

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TRI-MOR Standard Castable
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Heat treatment furnaces with special features

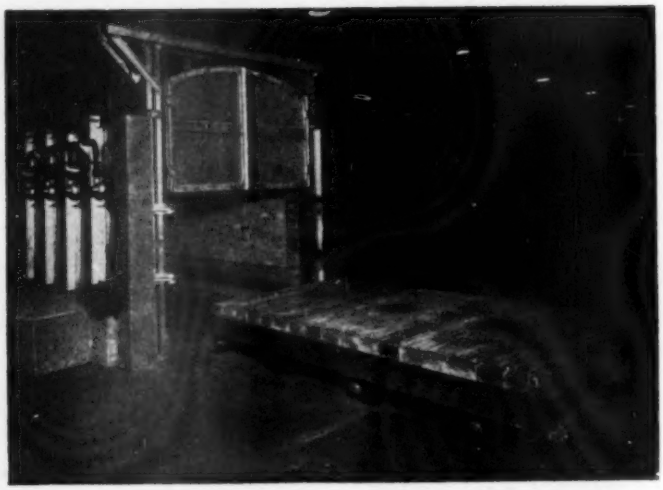

Photo by courtesy of Messrs. Lloyds (Burton) Limited

"Fulload" furnaces offer . . .

Maximum efficiency with minimum fuel consumption. The unique exhaust flue arrangement in the bogie hearth ensures speedy heating of the load, constant temperature in the furnace and the elimination of cold areas in the heating zone. We supply furnaces for all purposes and all fuels, batch type or continuous.

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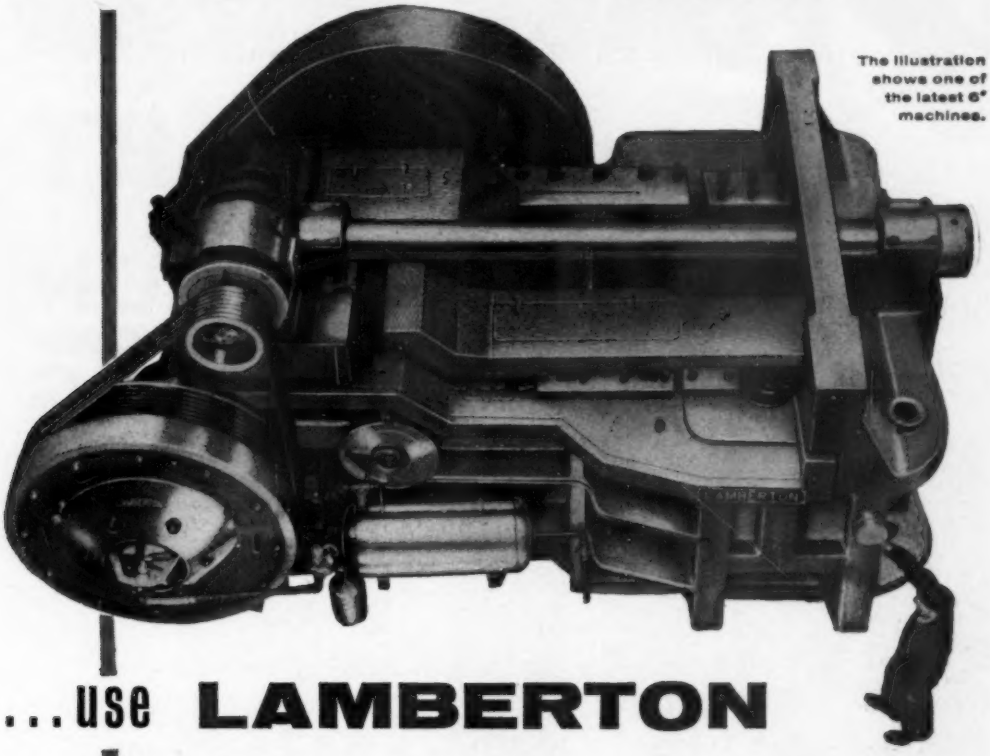

The illustration shows one of the latest 6" machines.

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HORIZONTAL FORGING and UPSETTING MACHINES

Automatic feeder can be provided as special attachment.

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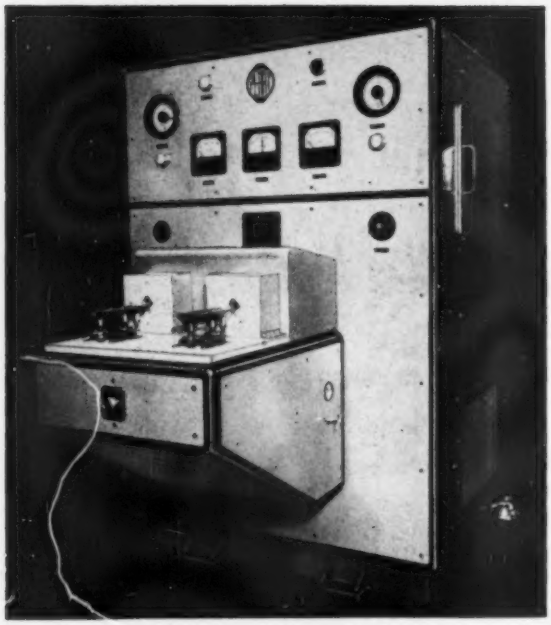

20Kw twin station automatic billet heater.

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★ *Absolute control over the heat input to the work*

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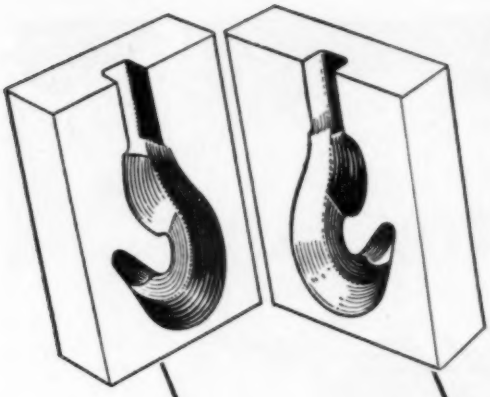
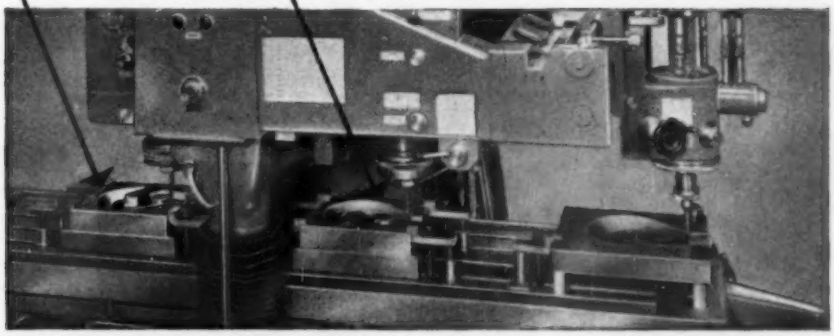


Both halves at once



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manipulator
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P.3359

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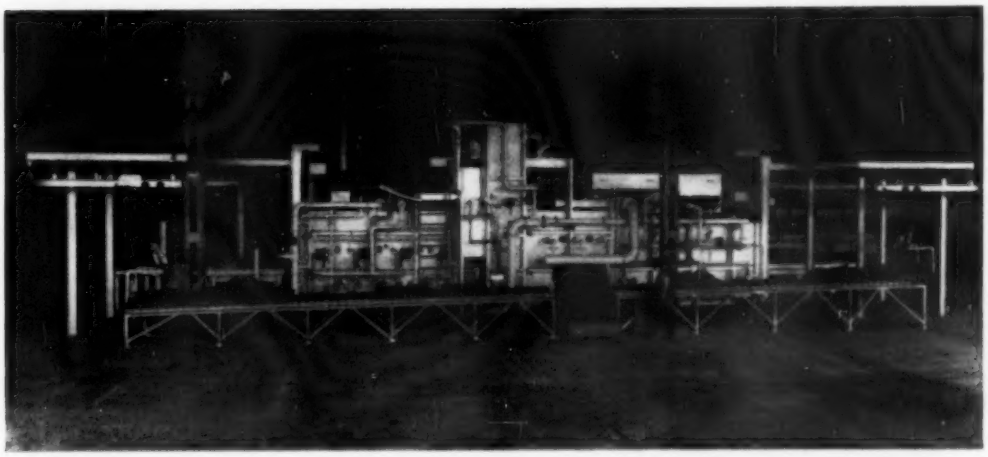

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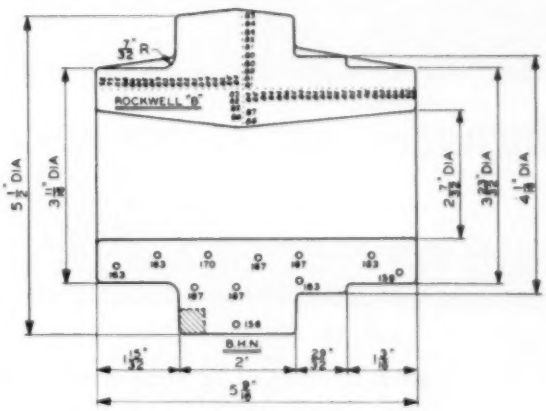

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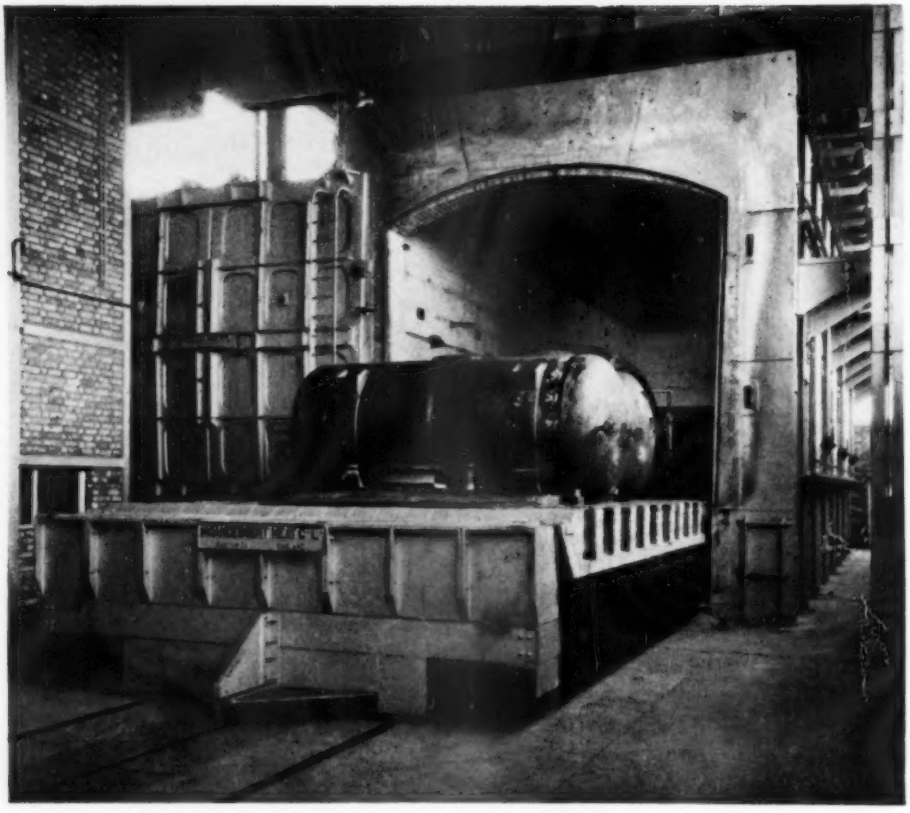

furnaces for stress relieving

INCANDESCENT have wide experience in building bogie hearth furnaces for many processes—stress relieving, annealing, heat treatment of forgings, mill rolls, etc. The furnace illustrated is installed at the works of Whessoe Ltd., Darlington, manufacturers of pressure vessels and chemical plant.

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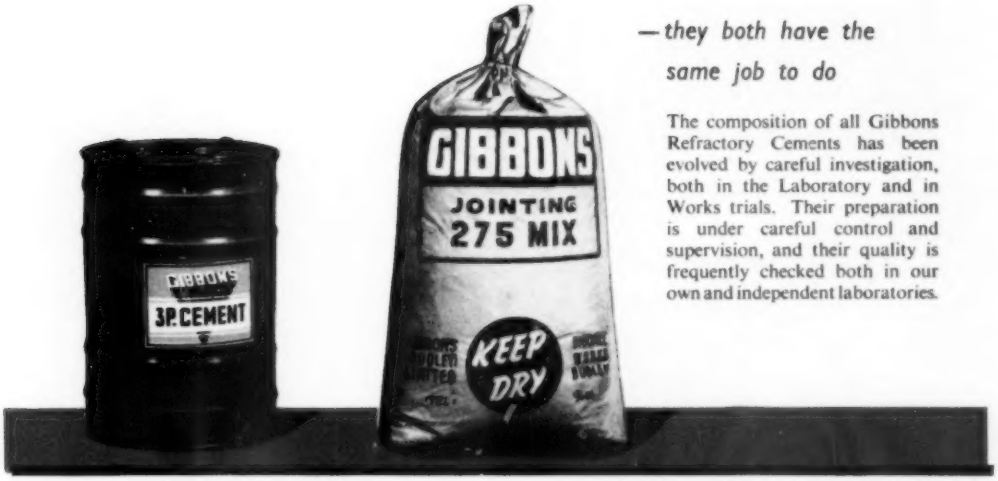

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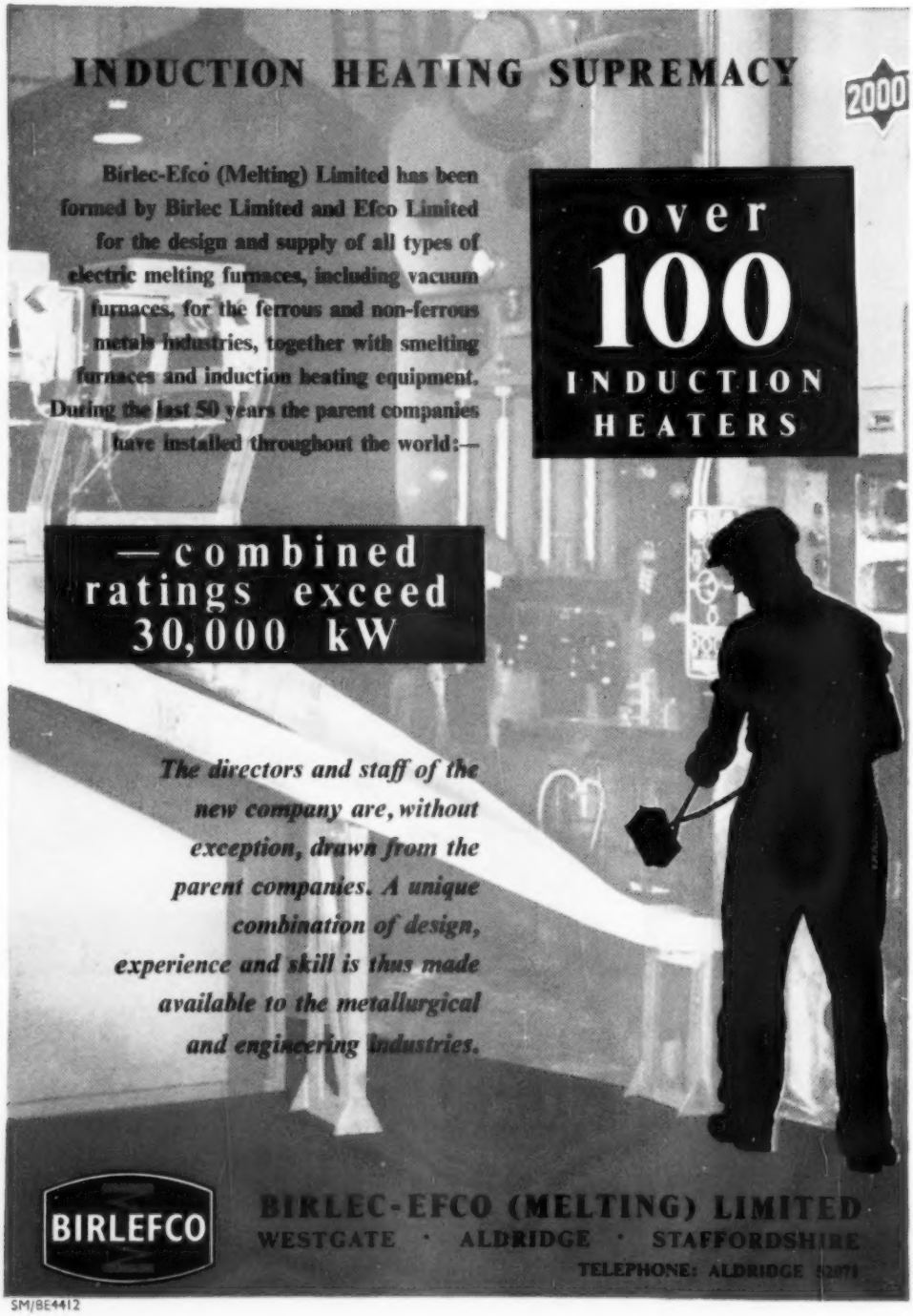
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2000
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September 1959

Vol 26, No 168

Metal treatment

and Drop Forging

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Rockets and research

THE spectacular success of the Russian moon rocket brings up once again the vexed question of how much of a country's total research capacity may safely be diverted to performing feats of national prestige. This is not, of course, to underestimate the value of such achievements for providing data which enable us to build up a more comprehensive picture of the solar system in which we live. In spite of fatuous remarks by some public figures, the 'conquest of space' appeals to the romantic in most of us, and much of the early pioneer rocket work was carried out by men who were directly inspired by such a romantic attitude. Nevertheless, it is quite certain that the colossal sums demanded by this type of project today would not be forthcoming solely for the purpose of increasing scientific knowledge and, whilst the propaganda stimulus is understandable, it is as well to consider this 'luxury' in relation to the total research effort of the countries concerned.

When we consider research in the U.K. in relation to research in America and Russia we come up against the obstacle of correlating figures based on quite different types of organization and any comparison can only be very approximate. However, some general figures may give a rough picture of the differences. In the U.K., the total number of qualified scientists in 1956 was 140,000, of whom 47,000 were engaged in research. Of this 47,000 more than half were employed on Government work or on Government-sponsored research. In the U.S., in 1954, there were 750,000 qualified scientists, of whom 157,000 were doing research. Only one-third of the research men were on Government work. When we turn to Russia, we are hampered by the difficulty of obtaining facts, but according to J. H. Hollomon in *Metal Progress*, 1959, 120, it seems possible that there were, in 1954, 700,000 scientists in Russia, of whom 100,000 were engaged in research and development. Moreover, it must be noted that the rate of training is increasing more rapidly in Russia than anywhere else, and in 1954 the number of scientists who completed training was 70,000 as compared with 36,000 in the U.S. and 11,000 in the U.K.

It is obvious that the relative sizes of population of the different countries determines that the research capacity of the U.K. is considerably less in total, but this does not alter the fact that only a small proportion is left for the industrial research on which the life of this country depends. Research projects do not diminish their demands on money and trained personnel in relation to the size of the population of the country in which they happen to be carried out. It is true that many industrial developments have resulted from research initially carried out on a Government programme, but in the field of metallurgy, for example, it is newer metals which have been favoured and not processes which find daily use in the bulk of Britain's industrial life.

Foreign competition should show us that it is in our traditional industries where our research effort has to be applied if we are to maintain our place in the world. This aspect of research may be less glamorous, and have less propaganda value than moon rockets, but it is not only more vital to our standard of living but is also within our economic possibilities. It will be a sad outlook for this country's future if we allow more of our already strained resources to be committed to keeping up with the propaganda race.

Japan's forging industry

ACCORDING TO the report 'Japan's iron and steel industry, 1959,' published by the Tokyo Foreign Service, Japan has 33 steel-forging companies with 36 works which are forging their own manufactured steel. There are also some 300 other forging works which manufacture steel forgings from materials purchased from other steel makers. The annual forging capacity is about 250,000 metric tons for each type of works. Forged steel has, like cast steel, a wide utilization. It is mostly used for parts for vessels, rolling stock, and machines. Shapes and kinds are numerous. Forged steel is used preferably for parts which require high physical, and chemical properties of high precision as well as for simple articles required in mass quantity.

TABLE I *Output of forged steel and consumption by industries in Japan*

Works which forge their own steel
(In metric tons as-forged)

Consumption	Fiscal 1957		1958 (Apr.-Sept.)	
	Output	%	Output	%
Ship and marine ..	63,988	34	14,092	22
Railways	33,694	18	14,520	23
Rolling mill and steel plant	33,484	18	11,549	18
General industrial machinery	19,894	10	7,292	12
Electrical machinery and equipment ..	15,467	8	4,567	7
Automobiles	8,352	4	3,389	6
Chemical industry equipment	6,497	3	1,723	3
Others..	9,003	5	5,579	9
Total	190,379	100	61,611	100

TABLE II *Output of forged steel and consumption by industries*

Works which forge steel bought from other companies
(In metric tons as-forged)

Consumption	Fiscal 1957		1958 (Apr.-Sept.)	
	Output	%	Output	%
Automobiles and bicycles	50,461	29	22,854	30
Machines	44,146	25	18,067	24
Ship and marine ..	26,838	15	9,465	13
Road and building construction	9,245	5	5,052	7
Railways	15,390	9	4,700	6
Others..	28,771	17	15,270	20
Total	174,851	100	75,403	100

TABLE III *Principal forging equipment*

Equipment	Number of stands
Forging press:	
Over 5,000 tons capacity	5
1,000—5,000 tons capacity	42
Below 1,000 tons capacity	48
Air-drop hammer	193
Steam-drop hammer..	66
Board-drop hammer	17
Forging roll	2
Forging machine	3
Forging manipulator..	11

Output of forged steel and its consumption by industries are shown in Tables I and II covering fiscal 1957 and the first half (April-September) of fiscal 1958:

Principal equipment for steel forging as of December, 1957, is shown in Table III:

These figures are from the forging companies which forge their own manufactured steel. The equipments of other forging companies (forge steel which is bought outside the company) are 577 stands of board drop hammer, 322 stands of air drop hammer, and 261 stands of other hammers, 18 stands of forging machines and six stands of forging rolls.

The forging equipment was modernized in 1958 by expanding heat-treating facilities, the use of modern auxiliary equipment for forging (manipulator, etc.), building of vacuum degasification machines and introduction of large-type machine tools.

The vacuum degassing method is now applied to forging 60-ton steel ingots with satisfactory results. Application will be extended to 80- to 100-ton ingots. A joint technical tie-up between the makers and the German Bochumer Verein and Leybold Companies was done so that forging of such large ingots would be possible.

This method has enabled makers to successfully turn out large-size ingots, especially fitted for shafts, electric power plants and shipbuilding as well as for large-size rolls of roll mills.

Among the various kinds of forged-steel products finding their way to overseas markets are rolling-stock shafts, anticlippers, large-type steel tubes for power plants, forged-steel rolls for rolling operation, die blocks and different kinds of machine tools.

Application of electron microscopy

Initial stages of fatigue in carbon steels

MIRKO KLESNIL and PREMYSL RYS

This article is the sixth of a series to illustrate the role of the electron microscope in various metallurgical investigations. The present work, which was reported in Hutnické Listy, 1958, (12), *describes research on plastic deformation and fatigue in carbon steels carried out by the Laboratory for the Study of Metals, Brno.*

THE TERM, FATIGUE OF METALS, indicates a process which progresses gradually from the moment when the component of the structure is first subjected to the oscillating stress. In order to acquire a deeper knowledge of the course of this complex process, which is the prerequisite for better utilization of the material under fatigue loading, it is clearly essential to trace the fatigue phenomena from the initial stages of the application of stress.

In the present work, which forms part of a research project of the Laboratory for the Study of the Properties of Metals, Brno, information is presented on the development of plastic deformation and fatigue phenomena in carbon steels with a low and medium carbon content in the primary phases of the application of fluctuating stresses.

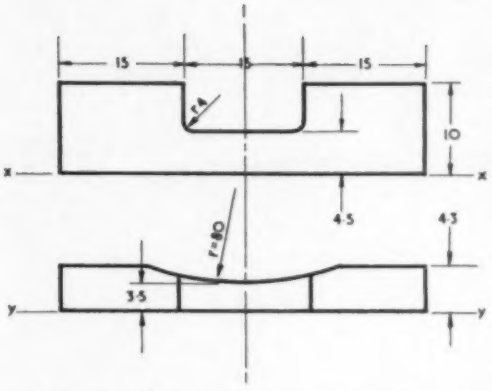
Review of previous work

The results of previous research work on the formation of fatigue failures of the cohesion of metals and alloys show agreement that these processes have their origin in the plastic deformation of the grains and of their boundaries. Likewise, there is no divergence of opinions on the expansion of fatigue cracks. At the end of the cracks which are formed in certain grains, high local stresses are set up by the notch effect, which leads to their expansion and, along the grain boundaries, to weakening of the load-carrying section and to ultimate breakdown of the cohesion. Opinions on the real nature and the moment of formation of the primary submicroscopic cracks, which represent the first phase of fatigue failure of materials, are up to the present time the subject of wide discussion.

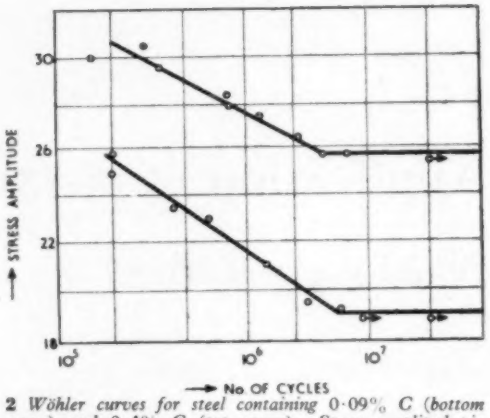
Gough,[1] Orowan[2] and Afanas'ev[3] maintain that local hardening of the material leads to breakdown of the cohesion at the moment when the stress in the hardened zones has reached the limiting value, since the ability of the material to undergo further hardening is completely exhausted. Wood and Segall,[7, 8] on the other hand, showed that in copper and alpha-brass hardening continues only up to a certain number of cycles. Thereafter increasing plastic deformation takes place without measurable hardening.

Oding[4] maintains that the formation of the first submicroscopic breakdowns is based on diffusion and piling up of vacancies forming in the slip planes due to interaction between the dislocations. His opinions appear to be in conflict with the results of the work of McCammon and Rosenberg,[9] who showed that fatigue phenomena of the same nature not only occur at normal temperatures, but also at 4·2°K. From this the conclusion may be drawn that a thermally activated process such as the diffusion of vacancies may play a part at higher temperatures, but is not an essential part of the breakdown, and the phenomenon of fatigue of the material is not dependent upon it.

Mott,[5] on the basis of an experiment of Forsyth,[14] who observed in aluminium and its alloys that the material was expelled in narrow strips from the slip bands, propounds the following mechanism for the formation of submicroscopic cracks. The expulsion



1 *Shape and dimensions of fatigue-test specimens*



2 *Wöhler curves for steel containing 0·09% C (bottom curve) and 0·4% C (top curve). Stress amplitude in kg./mm.²*

of the narrow strips of material and the formation of the submicroscopic cavities are brought about by the oscillating motion of the dislocation having a non-zero component of the 'Burgers' vector in the direction of the normal to the surface of the specimen; under oscillating loads this dislocation will run between two slip planes which are associated with oblique slip.

The theory of Cottrell and Hull[6] is similar; it also evades the question of the formation of cavities within the specimen, and explains the wrinkling of the surface by a slip in the two slip systems. Each cycle of applied stress produces elevations and depressions on the surface which are equal in height to the steps formed by the slip caused by the load factor which is equal to the stress. The depressions are, in the opinion of the authors, the first step towards the formation of the cracks.

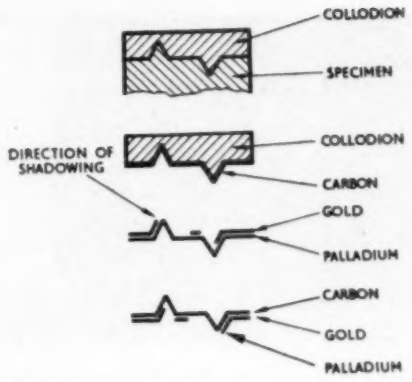
Fatigue processes have been investigated with the use of X-ray diffraction[8, 10, 11] and metallographically[6, 12, 13] for the most part on pure non-ferrous metals and their alloys. Noteworthy is the comprehensive work of Möller and Hempel,[11] who define the fatigue limit in low-carbon steel as the deformation limit, at which there is a change from homogeneous to inhomogeneous slip, leading to fragmentation of the grains. They suggest that the formation of the cracks occurs on the 'French' band; thus they attempt to give it physical expression. Metallographic examination has shown that plastic deformation under oscillating loads is much more localized to distinct zones of the grain than under load from a static force, and that the degree of localization is dependent on the frequency of application of the load. Apart from this, the relief forming on the polished surface of a specimen is completely distinct in its formation from the relief under static loading.

By means of the electron microscope it has been possible to obtain many more impressions of the structure of the zones of plastic deformation, the details of which lie outside the resolution powers of an optical microscope. Wewer and Hempel[12] observed changes in the structure of the surface of low-carbon steel specimens with the use of unshadowed replicas prepared from synthetic substances. But for correct interpretation of the phenomena which are considered to be submicroscopic cracks, their spatial representation is lacking. Observations of changes in the surface of specimens of aluminium and alpha-brass, subject to loading by fluctuating stresses, were published by Craig, Hunter and Fricke.[16, 17]

The primary microscopic cracks in the grains may be identified by the use of an optical microscope only by polishing the surface relief. Thompson,[15] in a study of fatigue phenomena in copper, after polishing at the starting points of the slip bands discovered traces of those which are called 'persistent slip bands.' Their depth after the application of stresses for 25% of the total period of cyclic loading required to produce fracture already amounted to 30 microns. After further polishing no new persistent slip bands were discovered. Thompson's persistent slip bands, as follows from the work hitherto carried out by us, may with the greatest probability be identified with the zones containing microcracks, which are formed in the grains of fatigue-stressed material even after a comparatively short period.

Experimental details

The specimens were prepared from carbon steels containing 0·09% and 0·4% C, and their shape is shown in fig. 1. Before being subjected to the oscillating load they were heated for a period of 1 h.



3 *Method of preparation of the two-stage carbon replicas and their shadowing*

at a temperature of 600° C. *in vacuo* to completely remove any internal stress which might have arisen during working. Cooling was carried out in the furnace over a period of 12 h.

The fatigue experiments were conducted on a symmetrical load, repeated flexure apparatus designed at the Laboratory for the Study of the Properties of Metals in Brno, with a frequency of 500 c./min., at normal temperature, without cooling the specimens. In order to determine the fatigue limit the number of cycles chosen was 10^7; the Wöhler curves for both types of steel are given in fig. 2. The fatigue limit of the steel with a 0·09% C content is $\pm$ 19 kg./mm.^{2} and of the steel with a 0·4% C content $\pm$ 25·5 kg./mm.^{2}

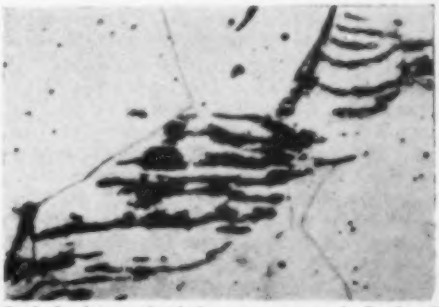
Our experiments consisted of the application of stresses to mechanically polished and chemically etched specimens, and of their metallographic examination at various phases of the application of the stress. In order to distinguish the structure of characteristic slip bands use was made of an electron microscope and two-stage carbon replicas, shadowed with gold and palladium. In the first place a primary collodion replica (fig. 3 (a)) was taken from the specimen, and a layer of carbon was applied *in vacuo* to the relief which had been so taken (fig. 3 (b)). After dissolving the collodion in amyl acetate, the carbon replica was shadowed either positively to determine the details on the level of the polished surface of the specimen (fig. 3 (c)), or negatively to determine the structure below the level of the surface of the specimen (fig. 3 (d)). The primary replicas were taken from the planes y—y (fig. 1). In figs. 9 to 12 the scale shown in the figures corresponds to a real length of 1 micron.

Results of metallographic examination

During fatigue loading, characteristic changes in the structure occur after a certain number of cycles, which grows with the decrease in the amplitude of the fluctuating component of the stress. On the surface of a specimen which is intended for metallographic examination with an optical microscope, in certain grains we observe characteristic bands which are normally referred to as slip bands. These bands are a manifestation of the oscillating plastic

4 *Optical micrograph of the slip bands in steel containing 0·4% C after 5 × 10^5 cycles at an oscillating load of 25·5 kg./mm.^{2}* × 400

5 *Optical micrograph of the slip bands in steel containing 0·09% C after 5 × 10^5 cycles at an oscillating load of 21·5 kg./mm.^{2}* × 800



6 *Optical micrograph of the same area as in* **5** *after 2·5 × 10^6 cycles at an oscillating load of 21·5 kg./mm.^{2}* × 800

7 *Optical micrograph of the slip bands during static loading* (σ = 30 kg./mm.^{2}) × 1,000

8 *Optical micrograph of the slip bands of steel containing* 0·09% *C after* 5 × 10^4 *cycles at an oscillating load of* 21·5 *kg./mm.*^{2} × 1,600

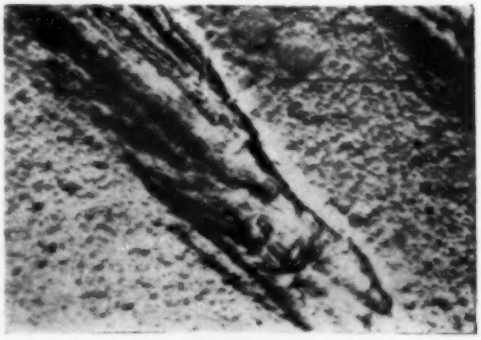
deformation and make their appearance first of all in the grains of ferrite, the orientation of the lattice of which is favourable to the advance of the stress which causes the plastic deformation (fig. 4).

In the initial stage the slip bands are limited by the boundaries of the ferritic grains; during subsequent loading they may pass through them into the neighbouring grain, if this is favourably oriented. In the majority of instances the development of these slip bands is obstructed by the pearlitic grains in steel with a medium C content. During further loading, in some of the slip bands cracks appear which extend relatively easily within the ferritic matrix and even across the grain boundaries, but for the most part their passage is obstructed by the grains of pearlite which are present in the structure. For this reason the intensity of the plastic deformation at the fatigue limit in the ferritic grains increases with the rise in the carbon content, which has been confirmed by metallographic study and by X-ray diffraction.[18]

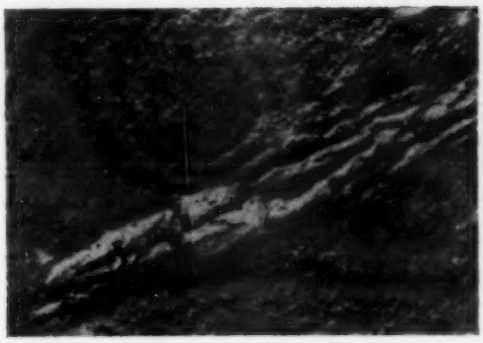
The initial stages of the formation of the slip bands and of the cracks forming within them were identical in both the steels examined. After a relatively small number of load cycles these slip bands are narrow (fig. 5); as the number of cycles increases for the most part they become markedly wider (fig. 6). The widening of the slip bands takes place only up to a certain number of cycles, and its extent varies in the individual, plastically deformed grains. Metallographic observation has shown that this process is completed at the moment when, as the submicroscopic cracks unite, a microscopic crack is formed of the same order of size as the dimensions of the grains. This fact, and likewise their different appearance by comparison with the appearance of the slip bands occurring during static loading (fig. 7), indicate that both the characteristic plastic deformation, as well as its manifestation, *i.e.* the slip bands, are quite different under oscillating loads from those under static loads.

Concerning the peculiar structure of the fatigue slip bands a metallographic photograph, even at high magnifications on an optical microscope, provides no closer information (fig. 8). In order to obtain more exact information on the nature of these so-called slip bands, we conducted a detailed study of them by means of an electron microscope. On the unshadowed carbon replica (fig. 9 (a)) it may be seen that under oscillating loads a slip band is formed by an area with a basic plane of expelled material. The slip band, therefore, does not in this instance represent an area of the grain in which movement of one part towards another has taken place, such as occurs under static load. Carbon replicas, it is true, give sufficiently high contrast at a suitable angle of application, provided there is variation in the thickness of the carbon layer under observation, but for a more accurate interpretation, and especially in order to obtain an impression of the dimensions of these deformations it is advantageous to use shadowed carbon replicas.

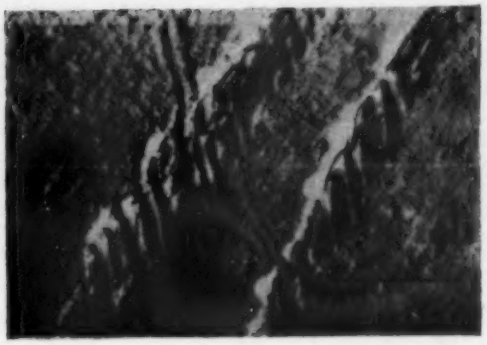
The further group of specimens (figs. 9 (b) to 10 (b)) were obtained from positively shadowed carbon replicas (see fig. 3 (c)). In fig. 9 (b) we see



9 (*a*)



9 (b)

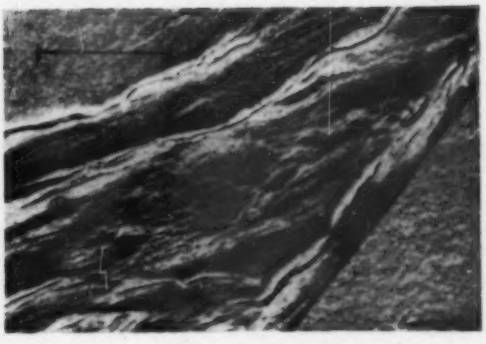


9 (c)

9 (d)

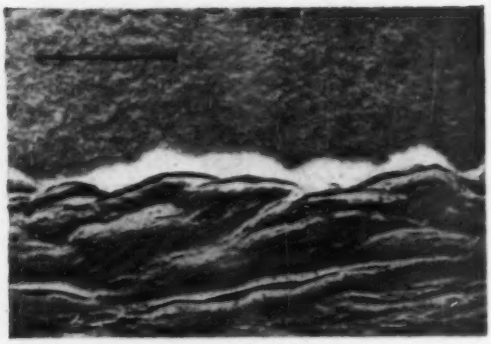
9 (e)

9 (f)

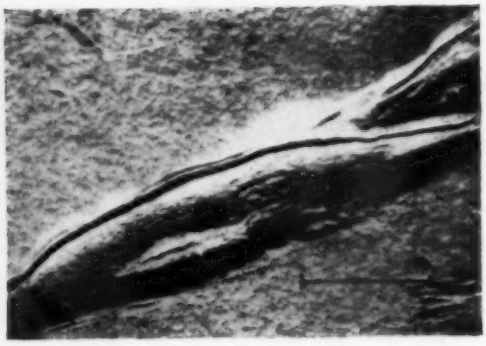


9 (g)

9 *Electron micrographs of the slip bands in steel containing 0·4% C after 10^5 cycles at the fatigue limit of 25·5 kg./mm.², (a) unshadowed two-stage carbon replica, (b), (c), (d), (e), (f) and (g) two-stage carbon replicas positively shadowed with gold and palladium, (a)—(d) show the course of slip bands in the ferritic matrix and (e) the relief formed by oscillating plastic deformation in two neighbouring ferritic grains with different orientations, (f) shows the transition from a zone of fine plastic deformation to a clearly marked slip band with sub-microscopic breakdowns and (g) an area of plastic deformation ending on the boundary of a ferritic grain.* *Scale shown = 1μ = approx.* × 20,000



10 (*a*)



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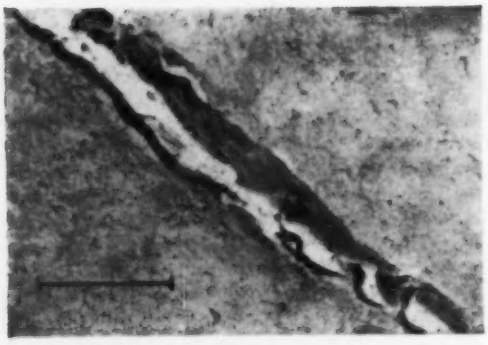
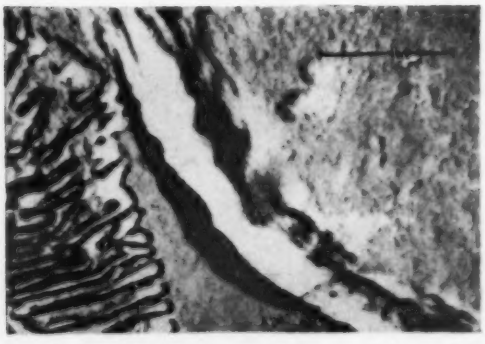
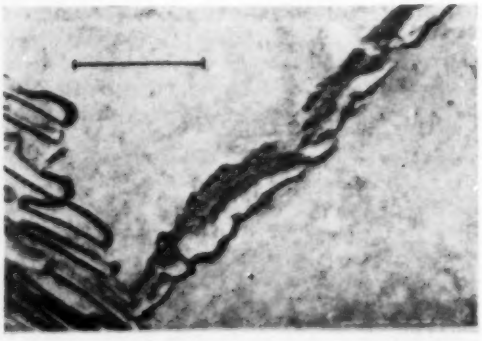


11 (*a*)

11 (*b*)



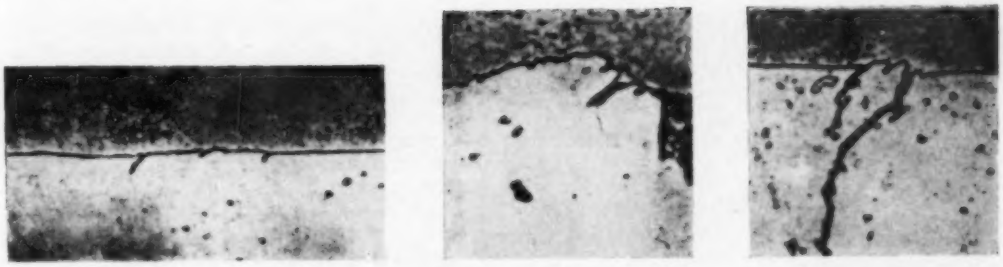
11 (*c*)



11 (*d*)

10 *Electron micrographs of slip bands in steel containing 0·4% C after 10^4 load cycles at 6% above the fatigue limit (σ = 27·0 kg./mm.²). Two-stage carbon replicas, positively shadowed with gold and palladium, showing zones of plastic deformation with sub-microscopic cracks*

11 *Electron micrographs of slip bands in steel containing 0·4% C after 10^5 load cycles at the fatigue limit (σ = 25·5 kg./mm.²). Two-stage carbon replicas, negatively shadowed with gold and palladium: (a) and (b) show slip bands with a series of sub-microscopic cracks casting shadows; (c) a slip band with a crack running alongside a pearlitic grain; and (d) how the development of a slip band and of sub-microscopic cracks is obstructed by a pearlitic grain. Scale shown = 1μ = approx. × 20,000*



12 (a) 12 (b) 12 (c)

12 *Micrographs of the surface relief of a cross-section of steel containing 0·09% C after 2·5 × 10⁵ cycles at a load of 21·5 kg./mm.². (a) the initial stage of the formation of microscopic cracks in the surface relief; (b) marked surface relief with cracks forming; (c) microcracks in an advanced stage of development*

the termination of a slip band in the ferritic matrix. No regular relationship between the course of the whole of the slip band and the detail of its relief is apparent, and in the light of a study of an appreciable number of slip bands, we established that this is true in the majority of instances. Exceptionally, the relief of slip bands has the appearance shown in fig. 9 (c), in which we see two distinct slip bands running side by side, which are made up of a collection of fine elevations and depressions inclined at a constant angle to the course of the slip band. It is remarkable that two small ridges of this fine relief run from the left-hand slip band into the surrounding unbroken surface of the ferritic matrix lying alongside the slip band. This phenomenon may probably be considered to be the initial stage of the further expansion which is reached as the number of cycles is increased, and which was shown in figs. 5 and 6. In fig. 9 (d) is shown a slip band, which to a certain extent has the same character, but which is running along close to the pearlitic grain; on the right of it lies a fine slip line which has clearly been brought to a halt by the same grain.

In fig. 9 (e) we see a relatively fine relief extending over wide areas of two neighbouring ferritic grains. The varying orientation of the individual slip bands of the deformed area may be seen in both the ferritic grains. A similar area with a fine surface relief which gradually develops into a clearly defined slip band is shown in fig. 9 (f). In fig. 9 (g) we see a zone of plastic deformation which ends on the boundary of a ferritic grain; the surface of the neighbouring grain remains completely smooth, without any relief at all. In the zone of plastic deformation we observe clearly defined dark lines, lying in the shadowed areas, which show up as the lighter areas on the specimens shown. Similar, but in essence more clearly visible, lines are to be seen both in an area inside the relief caused by deformation, and also along its boundaries in fig. 10 (a). In fig. 10 (b) a prominent dark line follows the boundary of the slip band. On both these specimens the dark lines lie in the shadowed areas, *i.e.* at points which are situated relatively lower than the surrounding parts of the relief.

By the technique used hitherto for the preparation of replicas, it has been impossible to determine beyond doubt the true nature of the dark lines in the deformation reliefs which have been described above. It is indisputable that these dark lines represent points with a fundamentally greater thickness of the film of carbon in the path of the passage of the electrons. This thickening of the carbon replica along the path indicated may be brought about by the presence of surfaces lying at a fairly oblique angle or even at a right angle to the polished plane, as may be seen from the appearance of the cementite lamellae of the pearlitic grain in fig. 9 (d).

The phenomenon may, however, be promoted by the presence of an extremely small crack, which is reproduced on the primary collodion replica. By the application of carbon to the vertical surfaces of this relief which were formed by the print of the crack, once again a much thicker layer is formed in the path of the passage of the electrons during the actual examination. From fig. 10 (a) and especially from fig. 10 (b) it is clear that the dark line represents a formation which does not lie on the surface of the relief because it does not cast a shadow. It may therefore be assumed that these dark lines represent submicroscopic cracks. In order to confirm this opinion we made use of negatively shadowed carbon replicas (fig. 3 (d)). If the dark lines are submicroscopic cracks, then by this latter technique for the preparation of the specimens they must cast a shadow. This fact is clearly confirmed by fig. 11 (a) and by fig. 11 (b) at a higher magnification. In fig. 11 (c) we see a relatively distinct microscopic crack which runs in close proximity to the pearlitic grain along its boundaries. In this instance also the shadow cast by the crack is clear evidence.

In fig. 11 (d) the pearlitic grain halts the advance of the plastic deformation and, therefore, the expansion of the slip band also. One of the cracks of this deformation relief ends in a cementite lamella of the pearlitic grain.

It follows from the study which was carried out on the electron microscope by means of two different methods of shadowing the replicas, that the slip bands which are formed during the application of oscillating loads to carbon steel are zones of deformation relief with a very reticulated surface of alternating elevations and depressions, for the most part without any clearly defined mutual relationship in their orientation. At certain points of this relief, breakdown of the cohesion of the ferrite occurs, and submicroscopic cracks are also formed.

In order to verify this concept we prepared a metallographic specimen of a cross-section. In order that the surface of the steel which had been stressed should remain unbroken after the preparation of the polished specimen, a thin layer of copper was deposited galvanically. The polished specimens prepared in this way were investigated at high magnifications on an optical microscope.

In fig. 12 (a) is visible a very slight relief with a few microscopic cracks in their initial stages. An area of more distinct relief may be seen in fig. 12 (b) on the side of which initial cracks are forming. Another small crack, the development of which has advanced considerably, and which extends to a fairly considerable depth, is visible in fig. 12 (c). Cracks of these dimensions can no longer be investigated by means of an electron microscope, since the primary replica is largely distorted when it is being separated from the specimen and so an artefact is produced.

Analysis of results

Investigation of the relief forming on the surface of specimens during plastic deformation under oscillating loads showed that the slip lines are composed of a series of elevations and depressions, which differ from one another in their dimensions. This reticulated relief is completely different from the appearance of plastic deformation under static loading, where movement of one part of the grain towards another takes place, and characteristic step structures are formed. The observations are compatible with the results of other work, apart from the fact that the surface phenomena of plastic deformation on certain pure, non-ferrous metals have a noticeably different character.[14]

During further application of the oscillating load the first submicroscopic cracks are formed in the zones of plastic deformation—this always occurs in the depressions. In the ensuing phase these submicroscopic cracks are united as macroscopic cracks within the framework of a single grain. Their expansion into the material of the ferritic matrix takes place relatively easily, even although the grain boundaries restrict this expansion, especially if precipitates are present. The pearlite which is present to a large extent obstructs their growth. This fact is in complete agreement with the increase in the fatigue limit of the steel in relation to the carbon content.

On the basis of the observations, the entire fatigue process may be divided into three main stages: (a) The stage during which hardening of the material takes place; this is similar to the corresponding stage during static loading. This stage ends after a certain number of cycles, and its duration is dependent on the amplitude of the deformation; (b) the stage during which further hardening of the material is immeasurable (in this stage the slip bands which have formed expand, and the first submicroscopic cracks occur within them); and (c) the stage during which one or more microscopic cracks extend from grain to grain, and in the ultimate phase this action leads to destruction of the cohesion of the material.

By means of theories of dislocation it is possible to explain the hardening of metals which are subjected to fluctuating dislocations. At a given amplitude of the deformation increasingly higher external stresses are necessary for the formation of new dislocations, since at a given amplitude of the external stress the amplitude of the deformation falls.[18] From this it may be concluded that the number of fluctuating dislocations is reduced. During the experiments it was established that in this stage of application of the oscillating load, the relationship between the stress and the deformation represents a closed curve, so that during the course of a load cycle plastic deformation, proportional in extent to the width of the curve, takes place. This plastic deformation may be brought about by the residual dislocations which fluctuate in cycles under the influence of the external stress between those dislocations which became firmly established in the preceding stage.

Metallographic observation shows that in the zones of plastic deformation submicroscopic cracks are formed on the surface of a specimen. It is difficult to suppose that the oscillating action of dislocation without hardening, *i.e.* the piling up of dislocation,[19] could be the cause of local breakdown of the cohesion of the material. But it should be taken into consideration that, on the surface of a specimen, oscillating plastic deformation brings about the formation of a relief which is formed of elevations and depressions. The depressions, especially, act as notches accompanied by localized increase in the stress. This localized plastic deformation takes place in submicroscopic volumes of the material which are quite inconsiderable in

relation to the volume of the specimen which is subject to stress. For this reason it is impossible that a localized increase in the stress can be registered by an increase in the external stress at a given amplitude of the deformation or by X-ray diffraction. In the notches the first crack is then formed, which extends from the surface of the specimen, as our observations have shown.

It may be supposed that the majority of the time required for fatigue fracture is taken up by the expansion of the cracks and not their formation, as is shown by figs. 10 (a) and (b). In these specimens is recorded the state of the surface of the specimen after tens of thousands of cycles, *i.e.* after less than 1% of the total time required for rupture of the specimen (see fig. 2). The fatigue crack is formed as a direct result of the oscillating slip in the second period of the fatigue loading, which follows the first stage after only a relatively small number of cycles. Under the conditions of fatigue loading described, cracks are formed in the steels investigated on any given surface of the specimen. This fact was confirmed by metallographic analysis of specimens which were gradually ground down. It was thus established that by grinding away the surface layers of specimens at various phases of fatigue loading it is possible to increase the number of cycles required to rupture the specimen; this confirmed that rupture starts on the surface.

The third stage of the fatigue fracture was not studied in detail in this work. It was, however, established that the duration of this stage is dependent on the quantity of pearlite, the grains of which obstruct the extension of many of the cracks from one grain to another.

Conclusions

In this work the sequence of fatigue phenomena was observed in materials which find extensive application in engineering practice, namely steels with contents of 0·09% and 0·4% C. In particular, investigations were made of the phase of this process which is marked by plastic deformation without measurable hardening, and by the formation of the first submicroscopic breakdowns of the cohesion of the material.

The technique which was used for the production of the two-stage carbon replicas with positive and negative shadowing, made it possible to distinguish details of the relief lying not only on, but also below, the level of the surface of the specimen under observation.

It was established that the breakdown of the cohesion of the material takes place exclusively on its surface, probably as a consequence of the concentration of stresses in areas which are located at the lowest points in relation to the surface relief. It was shown that submicroscopic cracks exist in the zones of plastic deformation even after a period under load which is only one per cent of the time required for rupture of the specimen. It was found that the pearlitic grains to a large extent obstruct the extension of the slip lines and of the microscopic cracks; this is completely compatible with the increasing fatigue limit of steel in relation to the growth in the carbon content. This is evidence that the duration of fatigue process in the steels which were studied consists predominantly of the development of breakdowns in the cohesion of the material, and not of their actual formation.

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Zirconium

Although zirconium was once considered a rare metal it is in fact more abundant than many common metals, but its use has been restricted up to now by its cost. The demands of the atomic industry have in this case, as with so many other costly materials, brought the use of zirconium within commercial possibility and it may be useful to become familiar with the properties of material. This article, based on a review in ' La Métallurgie,' January, 1959, is the seventh of a series on the new metals appearing in ' Metal Treatment '

ZIRCONIUM was discovered in 1789 by the German chemist, Klaproth, but he and other chemists were not able to isolate the metal. This was the work of two Dutch chemists in 1914, Lely and Hambruger, who produced a small quantity of impure metal in the laboratory. The pure metal was produced in 1925 by two other Dutch chemists, Van Arkel and De Boer, using the method known as the iodide process. This method was long and costly and was not adaptable for large-scale production but it was the only method used commercially up to 1947, when Dr. W. J. Kroll adapted his process for the reduction of titanium-magnesium to the production of zirconium.

Formerly classified as a rare metal, zirconium is, in fact, as abundant in the earth's crust as is chromium and consequently comes, in this respect, above many of the common metals including copper, lead, zinc and nickel. Hafnium is always in association with zirconium, the principal source of which is zircon, and to a lesser degree baddeleyite, an oxide of zirconium.

Mining the ore

Commercial concentrations of zircon are almost exclusively found in beach sands, from which it is extracted as a co-product of ilmenite, rutile and monazite. A small amount is produced from treatment of pegmatite. Baddeleyite is likewise present in river gravels and beach sands, the only commercial deposits being found in Brazil.

Although zirconium ore is found in at least 15 different countries, the six principal producing countries are the United States, Japan, Australia, Brazil, French West Africa and India.

Speaking generally, world reserves of zirconium ore are immense, and in the U.S.A. itself they are estimated to be between 5 and 15 million tons. They are considered to be larger than any foreseeable demand over the next 100 years. Existing reserves in India are moderately estimated at 5 million tons and those in Brazil, baddeleyite, at 2 million tons.

The cost of mining zircon and baddeleyite is low as compared with that of transporting and selling the ore. These are the main factors which stand in the way of the development of a number of deposits in distant locations, although these deposits are as rich and even richer than those which are actually being exploited.

In the U.S.A. beach sands are usually mined by suction dredgers, Humphrey spiral pumps, and magnetic and electrostatic separators. Practically the same methods are used in Australia, where the subsoil, with a thickness going to as much as 10 metres (33 ft.), is removed by bulldozers or mechanical shovels. The black sand thus uncovered is the selectively extracted, either with hand shovels or mechanically with suitable equipment. At least two companies are working there with dredgers. Heavy ore concentrates are obtained by gravity separation methods by using vibrating tables and Humphrey spirals. Zircon Rutile Ltd. extract a high zircon content ore by flotation but other firms dry the whole of the concentrate and obtain a saleable product by electrostatic and electromagnetic methods. In India extraction is chiefly by hand.

Treatment in the mill consists of sieving, followed by passing over wind-tables and magnetic separators.

Prospects

The outlook for zirconium is undoubtedly very bright. The main problem which has prevented the establishment up to now of a flourishing zirconium industry has been the impossibility of making reductions in production costs of the pure metal. When the U.S. Bureau of Mines began producing the metal in 1945, by the iodide process in their laboratory at Albany (Oregon), the metal was available only in small amounts at the price of $70/lb. There is still an outlet for exceptionally pure metal at this price for highly specialized requirements in electronics and dental prosthesis.

In 1953, private industry began to be interested and the Carborundum Metals Co. Inc., of Akron (Ohio), quickly developed commercial production, the largest in the world, of zirconium with a low hafnium content. Prices soon fell and the price of commercial zirconium in lots of 500 lb. and over dropped to $14·40/lb., while the ingot price of zirconium with a low hafnium content was reduced from a figure of $33 to one of $23·07/lb.

The American Atomic Energy Commission announced that they had placed contracts for an annual supply of 1,000 tons of zirconium over a period of five years, at an average expenditure of $15 million a year, with three companies, namely, National Distillers Products Corporation; National Research Corporation; and Carborundum Metals Company. The three companies will offer the metal at about $6 to $7/lb. The pilot plant at the U.S. Bureau of Mines at Oregon is in process of reorganization so as to give an annual production of about 135 tons. The three companies are each building new plants, with an initial capacity of 750 tons, which is well above the quantity called for under their contracts, on the assumption that the excess capacity will be absorbed, when the time arrives, by commercial atomic plants and for industrial uses.

In the U.K., zirconium has been produced for a number of years past by Murex Ltd., and this company has contracts for the sale of its output of zirconium from its enlarged pilot plant for the coming three years.

The chief consumption of zirconium in the metallic state is at present in nuclear-energy projects, this being due to the excellent corrosion resistance of this metal and to the fact that it does not absorb the neutrons, necessary to maintain a chain reaction. The Atomic Energy Commission demands a zirconium free from hafnium, as hafnium captures the neutrons. A very pure metal can be obtained by means of a modification of the Kroll process, or, alternatively, a spongy zirconium of fairly high purity produced by this method can be refined by the Van Arkel iodide dissociation process so as to give a metal with an exceptionally high degree of purity.

These are not the only qualities of this metal, which is almost as strong as steel but much lighter. It is chemically inert; a good heat conductor; its melting point is high, as is its specific weight. For this reason it is a particularly good refractory material. Zirconium-bearing refractories are specially useful in equipment for melting aluminium, as they are not wetted by molten aluminium. Moreover, zircon sands are superior to silica sands because they give quicker cooling, since they are better heat conductors. Zirconium, once it is stabilized, can withstand temperatures of 2,535°C. and it is an excellent electricity conductor at temperatures above 1,500°C. These properties make it a good material for jet engine linings and for H.F. insulating inductors and resistance furnaces.

The real obstacle which prevents zirconium being a redoubtable competitor in trade markets is its price. This is a serious handicap, but, when a not too expensive continuous process of producing this metal has been discovered, its exceptional properties of corrosion resistance will undoubtedly create a large outlet for it.

Metallurgy of zirconium

One of the most important chemical properties of zirconium is its sensitivity to the action of oxygen and nitrogen. It becomes extremely brittle if exposed to these two gases at high temperature. Less than 1% of oxygen and nitrogen make the metal brittle and no de-oxidizing or denitriding agent has yet been produced industrially.

Zirconium production presents very great difficulties. To produce a malleable metal, the Kroll process is being used, followed by the De Boer purifying process.

The Kroll process can be summed up as being the action of magnesium on zirconium chloride followed by vacuum refining. The impure metal, subsequently treated with iodine, allows of an almost pure iodide being produced, which, by distillation *in vacuo*, results in the production of the pure metal (De Boer).

In actual practice, zirconium chloride is obtained by the chlorination of a pure carbide, which is itself produced in an arc furnace with a graphite crucible 2¾ in. thick. An electrode, 4 in. dia., supplies current at 80 V. 2,500 amp. Production is about 250 kg. per heat and electricity consumption is 12 kWh. per kilogram of carbide. Silicate added to the furnace reduces the silica to silicon monoxide. Chlorination of the carbide is much easier than that of the oxide and it is this which has led to the success of the Kroll process. Great difficulties have

recently been overcome in the production of pure metal by elimination of the chief impurity in zirconium, namely hafnium.

Another difficulty in the use of the metal has been in the preparation of alloys by duplex melting. This preparation was at first done in graphite crucibles but is now carried out in special cooled furnaces, made of copper with tungsten electrode. The arc made in an argon atmosphere is obtained at 20 V. 400 amp.

Zirconium, of which the surface has a very low figure of neutron absorption (0·0010), is a metal with good mechanical properties with good resistance to corrosion by most of the chemical agents.

Zirconium alloys

Table I gives the properties of zirconium-hafnium alloys, which are used in the atomic industry with a low hafnium content (less than 0·50%). Alloys of zirconium with titanium, aluminium, copper and tantalum are given in Tables II to V.

Aluminium, tantalum, nickel and beryllium increase the resistance of zirconium alloys to oxidation. On the contrary, the alloy of zirconium with 50% titanium already reacts at a temperature of 750°C. Speaking more precisely, oxygen increases and nitrogen reduces the resistance of zirconium to oxidation.

Zirconium alloys containing aluminium, tantalum and columbium (niobium) have been tested with a view to checking their resistance to oxidation under steam pressure at 315°C. Aluminium reduces the resistance to oxidation. On the contrary, tantalum and niobium have no appreciable effect on the resistance of zirconium to oxidation.

Zirconium-beryllium alloys

Difficulties in the use of zirconium alloys, particularly the difficulty of melting it in an inert gas atmosphere, together with the troubles connected with the resistance of the known alloys to oxidation, have shown how important are the zirconium-beryllium alloys, which seem particularly suited for resisting oxidation.

Moreover, both these metals offer excellent resistance to neutrons, which they absorb to a very small extent. This property, which is only surpassed by that of bismuth and lead, is the reason why these two metals are of considerable importance in the atomic industry, notably in the construction of nuclear reaction moderators.

Finally, certain alloys based on beryllium have with low weight greater mechanical strength than aluminium, magnesium, titanium and stainless steel. Unfortunately, melting of both metals, zirconium and beryllium, necessitates certain special precautions owing to the fact that an inert-gas atmosphere must be used.

TABLE I *Properties of zirconium-hafnium alloys*

Hf %	Elastic limit kg./mm.²	Max. strength kg./mm.²	El %	Annealing temperature °C.	Annealing time h.
0·10	18	32·3	1	925	1
0·12	28·8	43·6	1·5	925	1
0·16	17	29·5	14·5	900	2
0·50	24	40·8	5	900	2
2·34	12	25·1	26·5	700	16
2·34	29·5	62·7	5	900	16
2·88	16	28	19	900	2
6·20	18·5	34·3	12	700	16
8·20	19·5	35·4	3·57	925	1

TABLE II *Properties of zirconium-titanium alloys*

Ti %	Elastic limit kg./mm.²	Max. strength kg./mm.²	El %	Annealing temperature °C.
3·2	29·5	60	17·3	700
3·2	58·75	91	7·2	400
15	52	64	1·5	
35	78	92	2·7	725
35	68	96	5·0	
50	79	98	1·5	725
50	62·6	91	1·0	
65	81	87·7	1·2	725
65	79·5	91	0·5	
85	55	71	1·0	900

TABLE III *Properties of zirconium-aluminium alloys*

Al %	Elastic limit kg./mm.²	Max. strength kg./mm.²	Elongation %	Annealing temperature °C.
1	44	57	5·6	725
2	66	78	3·3	725
4	79·2	86	12	725
1	51	70	8·7	as rolled
1·5	70·4	91	8	—
2	64	79·5	0·5	700

TABLE IV *Properties of zirconium-copper alloys*

Cu %	Elastic limit kg./mm.²	Max. strength kg./mm.²	Elongation %	Annealing temperature °C.
0·9	56	90	4	as rolled
0·9	11	33·8	17·7	700
2·9	53	92	1·5	as rolled
2·9	16	40	9	700

TABLE V *Properties of zirconium-tantalum alloys*

Ta %	Elastic limit kg./mm.²	Max. strength kg./mm.²	Elongation %	Annealing temperature °C.
0·5	10	29·5	24	700
1	10	30·5	19	700
2·5	25	45·8	17	725
2·5	42	63·3	3	cold rolled
3	15·5	38·5	15	700
7·5	33·2	54	17·5	725
7·5	42	67	3·5	
12·5	28·7	78·8	2·5	cold rolled
17·5	65	97	0·8	
22·5	65·4	78·8	4	
27·5	70	75	3	as rolled

Beryllium copper

Part 1 Working, machining and joining

G. FITZGERALD-LEE, F.R.Econ.S., M.I.E.I., A.R.Ae.S.

Beryllium copper with its combination of good mechanical properties, electrical and thermal conductivity, and resistance to corrosion is a most useful alloy in such applications as springs and electrical contacts. The following article giving some of the general working properties will be concluded next month with a consideration of heat treatment procedures.

Beryllium copper, like most other alloys, is made by melting the constituent metals together in the required proportions and pouring into suitable moulds, which may be either of simple shapes to facilitate subsequent fabrication by, for example, rolling or drawing, or in the form of castings.

Copper-beryllium master alloys, with about 4% beryllium, are commercially available for addition to copper. Although about equal amounts of copper and master alloy are required for the making of beryllium copper, it is better to melt the copper first and to add the master alloy to it, allowing for a loss of about 0·1% beryllium. Alternatively there is a variety of proprietary compositions on the market in ingot form for the production of castings. A proportion of beryllium-copper scrap can also be incorporated with due discretion.

Care should be taken to avoid the introduction of hydrogen during melting. Although molten beryllium copper is not particularly prone to absorb gas, cases have occurred in which wet or oily scrap, or the addition of damp charcoal, has led to unsound castings. Charcoal is normally used during melting, but it should be dried and preferably added in the incandescent condition.

Barium chloride is a suitable flux for dealing with small scrap or turnings. Fluxes like cryolite—which can react with beryllium—should be avoided; not only do they lead to loss of beryllium but they may also evolve fumes which, laden with beryllium compounds, are detrimental to health.

Health precautions

Certain compounds of beryllium, notably the oxide, are toxic and therefore detrimental to the health if inhaled or otherwise absorbed into the body. Accidents have occurred among operatives and others handling fluorescent electric lamps, some of which are coated internally with beryllium compounds but none has ever been reported in connection with the manufacture or use of beryllium copper.

As a safety precaution, furnaces in which beryllium copper is melted should be hooded and provided with extractor fans, and adequate forced ventilation should be installed in the casting shops. The use of respirators is desirable in foundries and also when welding—especially if fluxes such as fluorides, which are likely to carry volatile beryllium compounds into the atmosphere—are used.

Once the metal has solidified, however, it is perfectly harmless, and beryllium copper can be handled in press shops, machine shops and heat treatment plants with complete safety—except for the possible formation of oxide dust during high-temperature heat treatment. In subsequent use the metal is innocuous.

Pouring and casting procedure

In order to minimize the loss of beryllium in melting, the melt should not be overheated. For normal purposes a pouring temperature of about 1,100°C. should not be exceeded, this being approximately 100°C. above the melting point of the alloy.

Ingots and billets for subsequent fabrication are normally poured into cast-iron moulds, and oily mould dressings of the type used for brass and similar alloys give satisfactory results. The molten metal is very fluid and, provided that the melt is not gassy, it sinks well during solidification, necessitating adequate feeding. Care should be taken

not to expose too long a stream of molten metal to the air during pouring, in order to avoid oxidation with consequent loss of beryllium and the possibility of entrapping dross in the casting.

Moulding procedure for the production of sand castings is essentially the same as that for other copper-base alloys and the same sands and core materials can be used. Good permeability is advantageous and, to avoid hot tears, cores should not be too strong. The moisture in green sand moulds should be restricted and for large castings skin drying is desirable. The metal should enter the moulds as quietly as possible through a choke followed by a gently widening gate to slow down the stream. A skim riser to collect dross should be located behind the choke. Such precautions are recommended because beryllium tends to form a tenacious film of refractory oxide on the surface of the molten alloy—like aluminium.

The pattern shrinkage allowance for beryllium copper is about the same as that for phosphor bronze or gunmetal and lies between about $\frac{1}{8}$ in. and $\frac{3}{16}$ in. per linear foot.

In addition to the normal sand-moulding technique, beryllium copper castings can be produced by investment casting or shell moulding, investment casting being particularly suitable for small objects of intricate design where good surface finish and close dimensional tolerances are required. Semi-permanent moulds of the ceramic or gypsum type can also be used.

In the production of dies and die inserts for plastics it is not unusual to apply pressure during solidification, by means of either compressed air or a steel die—'hob'—inserted into the mould, in order to ensure the accurate reproduction of fine detail on the surface of the casting.

Although a few successful experiments have been reported on the die casting of beryllium copper, this is still largely in process of development for production on a commercial scale.

Hot working

Beryllium copper can be extruded and hot forged or pressed at temperatures between about 570°C. and 810°C., preferably towards the upper limit of this range. On the subject of hot rolling, information is still scanty, although commercially-produced sheet and strip has been broken down hot. For optimum response to subsequent precipitation hardening, however, hot rolling and other hot working operations should be carried out at temperatures between 750°C. and 810°C., and the metal should be cooled quickly, preferably by quenching. Such advice may not always be practicable, in which case a solution heat treatment should be applied at the end of the hot-working schedule.

Cold working

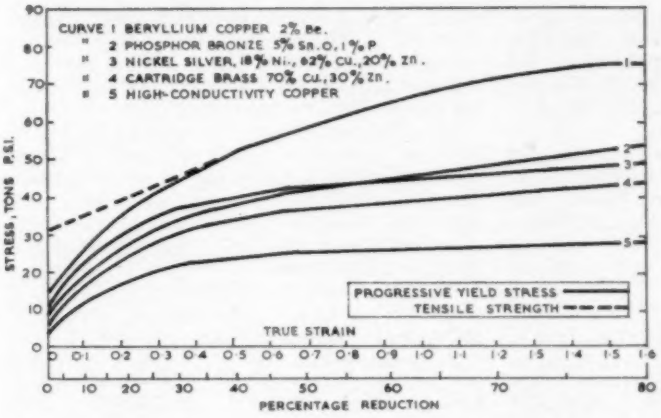
Cold-working processes are divided into two groups: those—like rolling and drawing—which are usually the responsibility of the manufacturer, and those—like press-forming operations—which are more the concern of the user. The same basic principles apply to both groups but they are considered here mainly from the point of view of the user.

Fig. 1 shows typical strain-hardening curves for annealed copper and several of its alloys, including normal beryllium copper in the solution-annealed condition, which lies wholly above its companions and shows a greater overall rise, indicating that beryllium copper is distinctly more difficult to deform than the others. It requires more power and more frequent annealing. Moreover, each anneal should, in theory, be carried out at the temperature of solution heat treatment, followed by quenching. This is not, however, mandatory for intermediate anneals, provided that the ready-to-finish anneal is so accomplished.

With reference to fig. 1, the true strain at any particular instant during the reduction of a sectional area of original size A_0, is defined as the minute change of area, $-dA$, at the instant in question divided by the sectional area, A, from which the change takes place. The actual true strain, obtained by integration, is therefore $\ln \frac{A_0}{A}$. Besides being more fundamental in concept, true strains have the great advantage over percentage reductions of being directly additive. Thus, if A_1 represents an intermediate stage in the reduction between A_0 and A, then $\ln \frac{A_0}{A_1} + \ln \frac{A_1}{A} = \ln \frac{A_0}{A}$, which is clearly the correct answer. In contrast, an attempt to add the corresponding percentage reductions leads to $100 \left(\frac{A_0 - A_1}{A_0} + \frac{A_1 - A}{A_1}\right)$ and not to the expected total reduction of $100 \left(\frac{A_0 - A}{A_0}\right)$.

A true strain of about 0·4, corresponding to about 33% reduction in sectional area, is generally found to be practicable between anneals. This degree of deformation, usually effected by cold rolling or drawing, brings the material into the hard temper, while the half hard and quarter hard grades correspond to true strains of about 0·2 (18% reduction) and 0·1 (10% reduction) respectively.

Since a total true strain of about 0·4 represents the practical limit of formability in each case, it is


CURVE 1 BERYLLIUM COPPER 2% Be.
" 2 PHOSPHOR BRONZE 5% Sn.,0.1% P
" 3 NICKEL SILVER, 18% Ni., 62% Cu., 20% Zn.
" 4 CARTRIDGE BRASS 70% Cu., 30% Zn.
" 5 HIGH-CONDUCTIVITY COPPER
STRESS, TONS P.S.I.
PROGRESSIVE YIELD STRESS
TENSILE STRENGTH
TRUE STRAIN
PERCENTAGE REDUCTION

1 *Typical strain-hardening curves for copper, beryllium copper, and some other copper alloys*

apparent that the formabilities of the four commercial tempers can be assessed roughly as:

Solution-annealed	.. 0·4—0	= 0·4
Quarter hard ..	.. 0·4—0·1	= 0·3
Half hard ..	.. 0·4—0·2	= 0·2
Hard	.. 0·4—0·4	= 0

These figures give a general indication of the relative degrees to which normal beryllium copper in the different commercial tempers can be bent, deep drawn or otherwise formed into useful shapes. Their significance is emphasized by the fact that smooth curves are obtained when the Erichsen values in Table I are plotted against them.

TABLE I *Typical Erichsen cupping-test results for normal beryllium copper prior to precipitation hardening*

Thickness of strip, in.	Solution-annealed	Depth of cup in mm.		
		¼ hard	½ hard	Hard
0·01	9	6	5	4
0·02	10·5	7·5	6·5	5
0·03	11	8·5	7·5	6
0·04	12	9·5	8	7
0·05	12	9·5	8	7
0·06	12·5	10	8·5	7·5
0·07	12·5	10	8·5	7·5

Low-beryllium copper behaves in much the same way and, for both types of alloy the fabricating qualities of the various tempers can be broadly described as: solution-annealed: for all applications involving severe deep drawing, bending or similar forming processes, may give slight burr on blanking; quarter hard: gives better spring properties and clearer blanking than solution-annealed material, and is almost as formable; half hard: good spring properties with formability about half that of solution-annealed material, can be bent around moderate radii and has improved blanking characteristics; hard: gives maximum strength, hardness and spring properties for parts which require little or no forming, good blanking properties.

Except in the form of finished products, normal beryllium copper is rarely supplied in the precipitation-hardened condition, for it is almost unworkable. In contrast, precipitation-hardened low-beryllium copper retains sufficient ductility to enable light forming operations to be carried out, and stock is sometimes supplied by manufacturers in the fully-hardened condition, thus eliminating the need for heat treatment by the purchaser. No great amount of deformation can be applied to such material, but in the manufacture of flat or moderately bent springs and similar articles, its use represents a considerable convenienee. On the other hand, it is more likely to carry locked-up stresses which are undesirable—especially from the point of view of resistance to fatigue.

Lubrication

Because of the comparatively high forces required to deform beryllium copper, proper attention to lubrication is necessary in all cold-working processes. Moreover, it is essential to ensure scrupulous cleanliness, and in particular to avoid oxide films which are abrasive in character and lead to unnecessary wear on the tools.

For light operations, a lubricant composed of soap and water is generally sufficient and has the advantage that it can easily be removed by washing. For severe deep drawing and similar purposes, however, it is desirable to use extreme pressure lubricants—although lard oil or a heavy soap paste with solid additions to promote film strength may

prove satisfactory. Lubricants containing appreciable quantities of sulphur tend to stain beryllium copper unless they are removed immediately after use, and in any case all lubricant should invariably be removed prior to heat treatment in order to avoid discoloration of the product.

For severe operations, it is possible to plate the stock with cadmium, which not only protects the surface from atmospheric oxidation but also acts as a lubricant and helps to reduce wear by cushioning the contact between tool and work-piece. The cadmium must be removed before heat treatment.

Punches and dies

The choice of materials for punches and dies in forming or cutting processes depends mainly on the length of run required. Carbide tools are long lasting whereas, in normal production, air-hardened steel can be used for moderately long runs or oil-hardened steel for shorter runs. Chromium-plated tools tend to reduce friction and chromium plating can be used to build up worn parts or to reinforce places at which wear is severe. With proper attention to cleanliness and lubrication, tool life for beryllium copper is about the same as for phosphor bronze, but much shorter than for brass.

For blanking and piercing, the cold-worked grades of beryllium copper give sharper edges with less burr than material in the solution-annealed condition. For the cold-worked grades a clearance between punch and die from 5% to 10% of the metal thickness is permissible, but for solution-annealed material the clearance should be held to a maximum of 6% of the metal thickness in order to minimize burr.

When blanking, the face of the punch is usually flat, while the die may be ground at a slight angle to reduce the pressure required. On the other hand, when piercing a small shear angle may with advantage be given to the punch. A cut edge should not approach the edge of the strip more closely than a distance equal to the thickness of the metal.

Deep drawing

The practical limits for the deep drawing of cups from the four commercial tempers of beryllium copper are approximately: solution annealed: ratio of blank diameter to cup diameter, 1·5; ratio of cup depth to cup diameter, 0·3; quarter hard: blank to cup, 1·3; cup depth to diameter, 0·3; half hard: 1·2 and 0·1; hard: 1 and 0. These figures are no more than a rough guide and, with this reservation, are applicable to both normal and low beryllium copper. They refer to conditions in which there is no ironing—in which the thickness of the cup wall remains the same as that of the blank. Moreover, to achieve them it is essential to ensure minimum friction by using smooth, well lubricated tools and no greater blankholder pressure than is necessary to prevent wrinkling. Such factors as metal thickness and radii on the punch and die may considerably modify the stated ratios.

If more severe drawing is required, it may be necessary to introduce one or more solution anneals into the schedule, so spaced that the total reductions between anneals do not exceed the stated figures. The metal should be cleaned both before and after annealing to avoid the possible formation of charred lubricant and to remove oxide scale. The material so treated will, of course, be in the solution-annealed condition and must subsequently be handled as such, whatever its original temper.

Solution-annealed and quarter-hard beryllium copper strip up to 0·04 in. in thickness can be bent through 90 deg. around a sharp radius without cracking. Nevertheless it is always desirable to use as generous a radius on the tool as design requirements permit. In the case of the half-hard and hard grades the direction of the bend relative to that of rolling becomes an important factor and the figures in Table II are a guide to the minimum radii required.

TABLE II *Minimum radii for bends in half-hard and hard grades of beryllium copper strip*

Thickness of strip (inches)	Minimum radius in inches for stated thickness of strip			
	0·01	0·02	0·03	0·04
HALF-HARD:				
Axis of bend parallel to direction of rolling	0·03	0·09	0·12	0·15
Axis of bend perpendicular to direction of rolling ..	0·03	0·06	0·1	0·13
HARD:				
Axis of bend parallel to direction of rolling	0·08	0·15	0·2	0·25
Axis of bend perpendicular to direction of rolling	0·05	0·09	0·13	0·15

Appreciable spring-back must be expected after bending beryllium copper and must be allowed for in designing tools. The degree of spring-back depends on a number of factors, including the severity of the bend and its direction relative to that of rolling, as well as the hardness and thickness of the strip. It can best be found by trial in any particular circumstances, but under normal conditions may amount from 5 deg. to 10 deg.

Spring-back can often be reduced by applying a moderate degree of coining either in the original forming process or in a subsequent operation. Close tolerances can, however, always be held by the

expedient of jigging during the final heat treatment. This is particularly valuable in the manufacture of spiral or helical springs from either wire or strip, and in all cases where the preservation of dimensional accuracy and reproducibility of performance of springs is of importance.

Cold hobbing is quite practicable with beryllium copper in the solution heat-treated condition. Tools must be robust and considerable power is necessary. Coining can be regarded as a comparatively mild form of hobbing, and the above remarks apply. It can be used in conjunction with other cold-forming processes and is, with some reservation, applicable to the work-hardened grades as well as to the solution-annealed material.

Although the above remarks refer more especially to the normal grade of beryllium copper the same principles are, in general, applicable to low-beryllium copper. This alloy is less hard and strong than the normal material and is consequently less exacting in the press shop. In the solution annealed or partially cold-worked condition its behaviour is comparable with that of phosphor bronze. Even in the precipitation-hardened condition low-beryllium copper retains sufficient ductility to enable light forming operations to be carried out—as previously mentioned.

Machinability

In the solution-annealed or partially cold-worked condition beryllium copper is comparable in machinability with other non-leaded copper-base alloys. The partially cold-worked grades are, on the whole, easier to machine than the solution heat-treated material, which tends to be 'sticky' and to drag under the tool. Tools of high-speed steel give satisfactory performance at moderate rates of cutting, but carbide-tipped tools can be used with advantage where rapidity of production is essential.

In the precipitation-hardened condition the normal alloy is usually too hard for ordinary cutting operations, but adjustments of size can be achieved by grinding. Precipitation-hardened low-beryllium copper can, however, be cut with carbide or diamond tools if necessary, although it is preferably machined before the final heat treatment.

Solution heat-treated stock is generally free from internal stress and is unlikely to distort during machining; this also applies to material in the precipitation hardened condition. Distortion due to internal stress is, in fact, only likely to arise in the case of the partially cold-worked grades.

While no clear recommendations concerning stress-relief heat treatment are as yet available, it appears that this could be accomplished at a temperature below that at which precipitation hardening becomes appreciable. From half to one hour at about 250°C. is suggested for purposes of trial.

Beryllium copper is often covered with a thin film of hard oxide and it is consequently desirable that preliminary cuts should be deep enough to reach well beneath the original surface—the film may alternatively be removed by pickling. Moreover, light cuts tend to work-harden the surface layers of beryllium copper, with consequent wear of tools. A depth of cut of at least 0·007 in. should be maintained wherever practicable. Rigid support of both work-piece and tool is essential and toolholders should be robust. Owing to its lower modulus of elasticity, an unsupported beryllium copper rod will deflect more under a given tool pressure than a steel rod of the same size.

A cutting fluid or lubricant should always be applied, an important function of which is to keep the metal cool. Without such precautions the surface layers may reach a temperature sufficient to cause partial precipitation hardening, with dire consequences to the life of the tool. Aqueous emulsions of the so-called 'soluble' oils have excellent cooling properties, while mineral and vegetable oils provide superior lubrication. In most cases the emulsions are to be preferred for beryllium copper. With either type of cutting fluid it must be remembered that the sulphur compounds which they often contain will stain beryllium copper if allowed to remain on the metal. Emulsions, unless removed with warm water, are apt to contract into droplets as they dry and, if sulphur is present, will lead to objectionable dark spots.

In general, tool designs as well as cutting speeds and feeds should follow recommendations for copper alloys which are strong and tough with homogeneous structures but which lack specifically free-machining qualities.

Joining processes

Beryllium copper can be successfully joined by welding, brazing or soldering, but all joining processes which involve the application of heat must be considered in relation to two fundamental properties of beryllium copper: its susceptibility to precipitation hardening and the formation of a thin but tenacious film of beryllium oxide on the surface when heated.

Welding should be carried out before the solution heat treatment, so that the whole of the product, including the weld zone, receives the same full heat treatment. Gas welding of beryllium copper is impracticable because of the rapid formation of beryllium oxide, which is difficult to remove by fluxing.

The carbon-arc method, using coated filler rods of beryllium copper, aluminium bronze or silicon bronze, is satisfactory providing that the melting is rapid and the rate of welding high. No flux is required in view of the blanketing effect of the

carbon arc gases. Metal-arc welding with coated beryllium copper or coated aluminium bronze electrodes gives good results. Suitable fluxes include ordinary brazing fluxes with admixtures of sodium carbonate and potassium bifluoride, and boric acid containing sodium fluoride. Spot, butt and seam welds in beryllium copper can also be made by the resistance-welding process, and good welds are made by the argon-arc process, without flux.

Brazing, and particularly silver soldering with alloys melting between 600° and 700°C., can be applied to beryllium copper, using the normal fluxes supplied with the solders. If, however, the surface is badly oxidized it should be cleaned mechanically before brazing, while solvent degreasing is recommended when specimens are heavily coated with grease.

Brazing and silver soldering should be effected quickly and are necessarily restricted to material in the solution heat-treated condition, since the temperature involved would destroy the effects of previous precipitation hardening. It is, however, an advantage that brazed or silver soldered articles can subsequently be hardened and strengthened by heat treatment at relatively low temperatures—even though response to heat treatment in the vicinity of the joint may be somewhat impaired.

Although the soft soldering of beryllium copper presents little difficulty, certain precautions are necessary if consistently good joints are to be obtained. It is of primary importance that the surfaces to be soldered should be free from beryllium oxide.

Since soft soldering must clearly be carried out after the final heat treatment—the melting point of solder being lower than the precipitation hardening temperature—it is evident that superficial oxide arising from both the high and low temperature heat treatments will be present on the surface unless special precautions are taken to remove it. A considerable amount of beryllium oxide may be formed during the solution heat treatment, nevertheless the precipitation-hardening treatment is much less detrimental from this point of view. It is, however, expedient to remove the oxide film *between* the two heat treatments since the second of these tends to make any beryllium oxide previously present harder to eliminate.

Where possible it is desirable to adopt abrasive or mechanical cleaning, using a file, emery paper or even metal polish, but in many cases—as in the mass production of minute components—this is hardly practicable; such articles should be pickled between the two heat treatments. In obstinate cases it may be necessary to repeat the pickling process after the final heat treatment; but it will generally be found that ordinary soft-soldering fluxes based on zinc chloride effectively combat any traces of oxide which may be reformed at this stage, as also do certain of the activated resin fluxes.

Acknowledgments

The author is indebted to the Copper Development Association for the information in this article.

High-temperature bearings steel

Use in jet engine

JET-ENGINE BEARINGS to operate at elevated temperatures are now made of a '99% native' alloy steel by the Bower Roller Bearing Division of Federal-Mogul-Bower Bearings Inc., of Detroit. The material requires no imported alloying components such as tungsten or cobalt. The new steel is identified as Bower '315' and produced by Republic Steel Corporation. Its principal alloying elements are 3% Ni, 1·5% Cr and 5% Mo.

The new bearing steel is being used on the 'hot' end of military jet engines. Operating temperatures range up to 500°F. (260°C.), about twice the temperature for bearings used in internal combustion engines for automobiles and trucks. While steel ordinarily softens permanently at elevated temperatures, Bower '315' recovers its original hardness when cooled to room temperature after heating as high as 1,000°F. (540°C.).

Heat treatment

Races and rollers for bearings are held at 1,000°F. as a final step of the heat treatment. Since this temperature exceeds the operating temperature for the bearings, '315' provides stability and prevents dimensional and hardness changes in the bearings in service. Races were tested at Bower by repeated heating and cooling between room temperature and 1,000°F. They showed no measurable change in hardness or dimensions measured to the nearest ten-thousandth of an inch. The coefficient of thermal expansion of '315' is close to that of the steels used for shafts and housings, according to the bearing manufacturer. This prevents difficulties which could be caused by differences in dimensional changes over the wide range of operating temperatures.

Produced as a low-carbon steel, '315' is readily machined. Case hardening after machining gives the bearing surface a hardness of 58—62 Rockwell C, with good wear resistance. On test stands, bearings made of the new alloy steel have operated without failure at temperatures up to 600°F. (315°C.). The tests include acceleration of the shaft from rest to 11,000 rev./min. in as little as 8 sec.

Hot working of alloy steels

Part 2 Essential factors for ductility

A. GUEUSSIER and R. CASTRO

This investigation into the hot ductile behaviour of alloy steels of varying compositions under varying conditions was described by the authors at the 1957 *autumn meeting of the Société Française de Métallurgie, Paris, and published in French in 'Revue de Métallurgie,'* 1958, **15,** (11). *The authors are both with Aciéries Electriques d'Ugine, Savoie, France, Mr. Castro being chief engineer, head of research. The first part of this study was given in the last issue and the third and concluding part will appear next month*

Influence of temperature

In the following examples, a certain number of curves will be found giving the variation in ductility as a function of temperature, other things being equal, of course. In actual practice, curves of widely differing form can be observed, but the more usual types correspond, however, either to a simple curve showing a maximum at high temperature or to a more complicated curve showing notably a minimum at a moderate temperature and a maximum at a very high temperature. However, in certain cases, curves have been found showing a number of minima and maxima.

The only phenomenon appearing in practically all cases for a range of alloy steels is a maximum at a very high temperature. Beyond this maximum the ductility of the steel decreases rapidly, as the temperature rises, to fall to zero, which corresponds to complete burning of the metal, that is to say to the formation in it of a continuous liquid film resulting from crossing the solidus line. The temperature of this maximum deformability, preceding the phenomenon of burning, fixes therefore the maximum temperature of working allowable beyond which the phenomena of overheating and burning begin to appear.

Influence of speed of deformation

The essential influence of this factor on the ductility of steels and alloy steels has already been put forward by Portevin, Prétet and de Lacombe[1] and has since been confirmed by various authorities. A few characteristic examples of it are given here.

Fig. 10 shows a mild steel (for analysis of all steels in this report see Tables I and II) forged down to 12 mm. ($\frac{1}{2}$ in.) square for which variations in hot ductility are shown as a function of temperature for varying speeds of deformation. These results have been obtained from hot tensile tests at the two available experimental speeds. Based on this figure, two ranges of temperatures can be defined, in which the metal behaves differently: at low temperatures the steel works better at high speed than at low speed, whereas, at high temperatures, the reverse phenomenon is observed. Similar curves can be obtained for low or medium alloyed steels.

Fig. 11 shows the results of similar tests, made under the same experimental conditions on an 18/8 type of stainless steel, forged down to $\frac{1}{2}$ in. square and quenched at 1,150°C. The results obtained were the same as in the previous example.

Investigations into the phenomenon of brittleness by means of bend tests with stainless steels are shown in the example given on fig. 12, which confirms that this brittleness is more pronounced in proportion as the speed of deformation is lower. Similar tests could be made for brittleness of the same kind taken at temperatures between 1,000 and 1,100°C., for mild steel and treated steels.

It will be noted that, in practice, the speed of deformation does not remain strictly constant neither during working under a given tool nor in

the course of a definite test. However, in both cases, the variations are not so excessive as to make it impossible to give an average speed of deformation for the whole of the operation. Experience shows, indeed, that within the range of speeds examined during these tests, this factor only intervenes in any noticeable way on the hot ductility of steels when it varies to a considerable extent, say, for example, in a ratio of 1 to 10.

TABLE I *Analysis of steels examined*

	C	Si	Mn	Ni	Cr	Other elements
A	0·080	0·50	0·61	10·30	17·70	—
B	0·055	0·40	0·32	0·27	17·20	—
C	0·054	0·23	0·38	0·26	0·15	Al = 0·014
D	0·136	0·56	0·65	10·30	18·00	—
E	0·068	0·58	0·54	9·80	17·90	—
F	0·765	0·31	0·47	0·20	4·10	W = 18·9 V = 1·08
G	0·110	0·53	0·45	10·20	17·90	—
H	0·105	0·20	0·40	0·14	6·20	Mo = 0·64
I	0·095	0·59	0·49	11·30	17·00	Ti = 0·31
J	0·141	0·35	0·36	0·14	6·40	Mo = 0·61
K	0·095	0·40	0·51	0·33	12·90	—
L	0·093	0·50	0·45	9·60	17·30	Ti = 0·56
M	0·062	0·49	0·50	12·20	17·00	—
N	0·060	0·54	0·51	12·10	17·00	B = 0·05% added
O	0·139	0·33	0·26	0·25	5·80	Mo = 0·54
P	0·054	0·41	0·37	12·30	17·00	—
Q	0·045	0·43	0·32	12·10	16·80	Mo = 1·42
R	0·050	0·40	0·39	12·40	17·10	Mo = 2·90
S	0·101	0·53	0·41	12·20	17·10	—
T	0·103	0·58	0·53	12·70	17·10	Ti = 0·13
U	0·102	0·59	0·54	12·10	17·10	Ti = 0·31
V	0·031	0·65	0·62	13·60	16·80	Mo = 2·35
W	0·033	0·66	0·69	13·70	17·20	Mo = 2·26 Ce = 0·3% added

TABLE II (A) *Analysis of steels*

	C	Si	Mn	Ni	Cr	Ti	V	Percentage of ferrite in the steel as cast
1	0·071	0·51	0·45	10·3	16·3	0·52	0·47	2/3
2	0·060	0·62	0·44	10·4	16·2	0·36	0·77	4/5
3	0·087	0·65	0·49	10·4	16·6	0·52	1·01	7/9
4	0·093	0·61	0·46	10·2	16·8	0·46	1·90	10/12
5	0·101	0·62	0·48	10·3	16·7	0·59	2·80	15/18
6	0·103	0·77	0·55	10·3	16·5	0·50	3·78	25/30
7	0·097	0·55	0·49	10·3	16·8	0·59	4·75	40/45
8	0·094	0·59	0·58	9·2	18·3	0·73	3·84	65/75
9	0·088	0·65	0·56	8·2	19·3	0·72	3·90	96/98

TABLE II (B) *Analysis of steels*

	C	Si	Mn	Ni	Cr
10	0·028	0·59	0·61	8·2	17·6
11	0·033	0·61	0·55	8·2	19·7
12	0·033	0·70	0·55	8·1	21·9
13	0·031	0·52	0·61	7·2	20·8
14	0·033	0·58	0·56	7·4	21·5
15	0·032	0·59	0·48	6·3	21·3
16	0·030	0·65	0·48	6·3	22·4

Influence of the deformation process

It has been stated that the ductility of a steel at a given temperature and for a given speed of deformation depends upon the method of working adopted (continuous or in successive passes). This phenomenon was also observed by G. Véry (private communication).

An example of this is shown in fig. 13, which shows bend tests carried out on a tungsten high-speed steel, 18/4/1 type, as cast. Test-pieces were taken from a 10 cwt. ingot, near the surface and parallel to it. Curve A shows the continuous deformation of the steel at high speed (20,000%/sec.) and shows the variation with temperature of the critical deflection f_c corresponding to the appearance of the first crack. Curve B was obtained under the same experimental conditions but with non-continuous deformation. The corresponding test-pieces were worked at high speed with a constant test temperature, by a succession of short passes. These passes correspond to definite elementary deformations of increases of 0·5 mm. (0·02 in.) bending deflection. A period of 2 minutes separates each individual deformation.

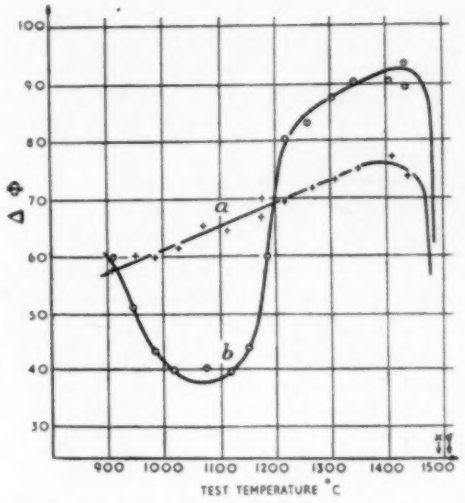
These tests were carried out in such a way that the total maintenance of temperature, at the end of working, was practically the same for both curves A and B, and was about 20 minutes. On the same diagram, curve C corresponds to the continuous deformation of the same steel at low speed (30%/sec.). Curve D is to curve C as curve B is to curve A; it represents a non-continuous deformation, each pass in this instance corresponding to an increase of 1 mm. (0·04 in.) in bending deflection carried out at low speed every 2 minutes.

This test shows that deformation in successive stages results, in this case, in an improvement in ductility, which is all the more marked as the speed of deformation increases. It will be noticed that the ductility of the steel worked at high speed, but in successive stages with temperature held during the intervening periods at test temperature, can approximate to the ductility obtained through continuous working at a much lower speed. Consequently, in order to represent experimentally what happens in practice under the hammer, one may be led to adopt either a test carried out in the non-continuous method with a high speed of deformation or a test with a continuous method at a much lower speed. This example shows how difficult it is to arrange for a test perfectly adapted to represent the behaviour of a steel under a given tool.

Influence of prior heat treatment

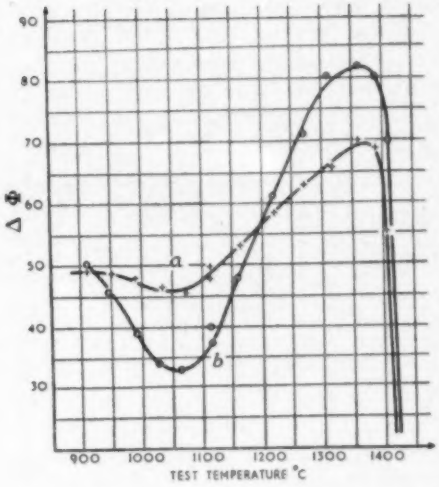
For certain steels the hot ductility depends essentially on the preliminary heat treatment.

The authors have already shown this phenomenon



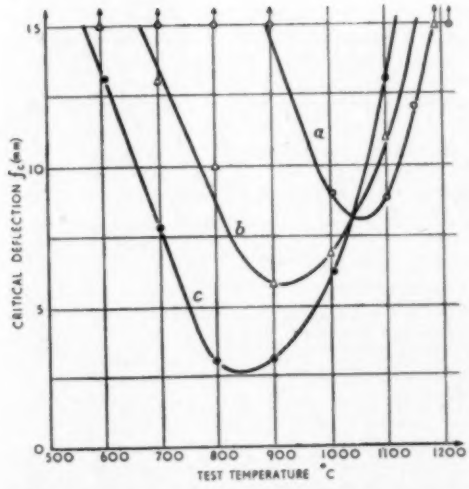
10 *Influence of speed of deformation. Steel C, mild steel, ½ in. square forged, air-tempered 950°C. for 20 min., hot-tensile tests*

Speed of deformation (a) *equal to* 40,000%/*s.* (b) = 500%/*s.*



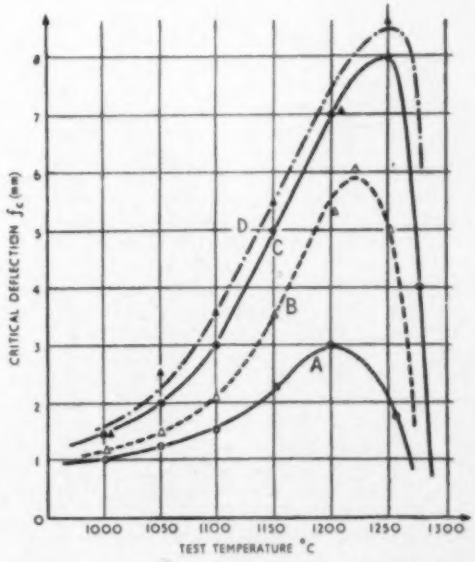
11 *Influence of speed of deformation. Steel D, 18/8 stainless, ½ in. square forged, water-quenched 1,150°C./1 h., hot-tensile tests*

Speed of deformation (a) *equal to* 40,000%/*s.*, (b) = 500%/*s.*



12 *Influence of speed of deformation. Steel E, 18/8 austenitic, ½ in. square, water-quenched 1,250°C./1 h., hot-bend tests. Test-piece 10 × 10 × 55 mm. U.F. notch*

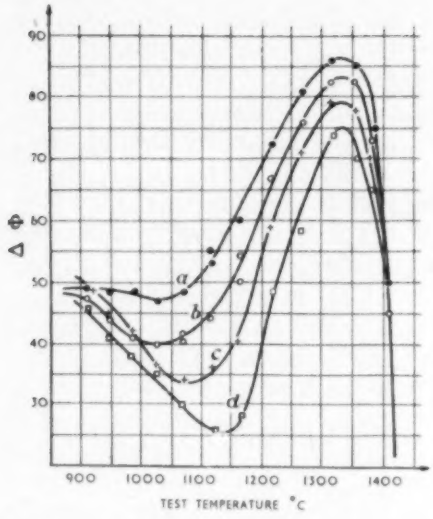
(a) 20,000%/*s.*, (b) 30%/*s.*, (c) 1%/*s.*



13 *Influence of method of working. Steel F, high-speed 18/4/1, as cast. Test-piece 6 × 10 × 30 mm., hot-bend tests*

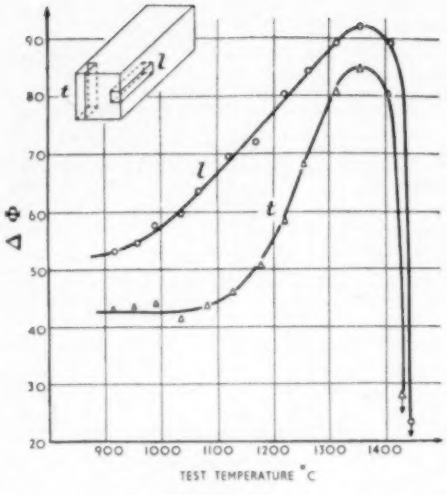
(a) (c) *Continuously worked,* (b) (d) *discontinuously worked*

(a) (b) 20,000%/*s.*, (c) (d) 30%/*s.*

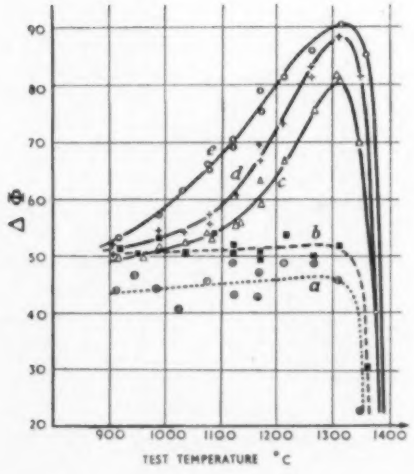


14 *Influence of preliminary heat treatment. Steel G,* 18/8 *austenitic,* ½ *in. square, hot-tensile tests,* 500%/*s.*

Water-quenched for 1 *h. at* (a) 950°*C.,* (b) 1,050°*C.,* (c) 1,150°*C.,* (d) 1,250°*C.*

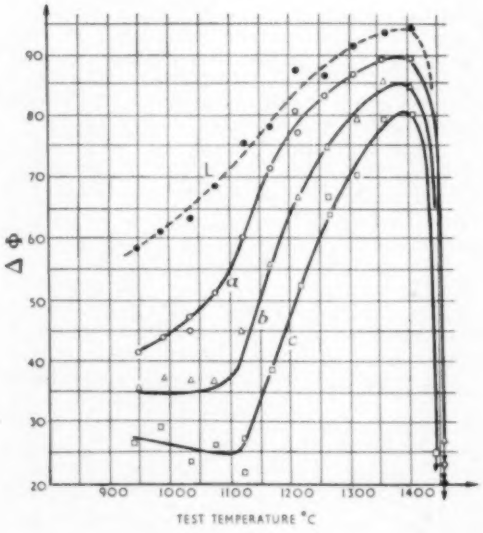


15 *Influence of method of taking test-pieces. Steel H,* 5% *Cr-*0·5% *Mo,* 3½*-in. square billet in annealed state* (900°*C.*/30 *min.*). *Test-piece* 3½ *in. for transverse study, calibrated* ¾ *in. dia.* ¾ *in. long, hot-tensile tests,* 250%/*s.*



16 *Influence of working on lengthwise ductility. Steel I,* 18/8, *Ti, air-hardened* 1,050°*C.*/1 *h., hot-tensile tests,* 500%/*s.*

(a) (b) *As cast,* (c) 2 *in. sq.,* (d) 1 *in. sq.,* (e) ½ *in. sq.*



17 *Influence of working in transverse ductility. Steel J,* 5% *Cr-*0·50% *Mo, annealed* (900°*C.*/30 *min.*), *hot-tensile test,* 250%/*s.*

Taken transversely (a) 7½ *in. sq.,* (b) 5½ *in. sq.,* (c) 3½ *in. sq.*
Taken lengthwise L ¾ *in. dia. forged*

for stainless steels of the 18/8 type in a previous report.[10] An example of this phenomenon is given in fig. 14 for a steel of the type used in the tensile tests of the present investigation. The variations in ductility of the steel with temperature have been determined after preliminary quenching treatments at rising temperatures. It will be seen that between 1,050 and 1,100°C. the brittleness becomes more pronounced in proportion as the initial quenching temperature gets higher.

It was also found that, at high working temperatures, the fine grain structure obtained at low initial quenching temperature resulted in better ductility than the large grain structure due to high quenching temperature.

This essential influence of first heat-treatment temperature on the ductility of austenitic steels must be taken into consideration when the steel is being worked at falling temperature (deformation in a number of passes). An initial heating at excessive temperature can lead to cracks at the end of rolling owing to the worsening in brittleness which it causes at lower temperature. Similar results have been observed by H. M. Soldan and C. R. Mayne[11] when they were studying the ductility of stainless steel tubes for superheated steam, and of their weldings.

Influence of method of taking test-pieces

There is evidence to show that there is a ' transverse ' hot phenomenon, similar to the well-known cold ' transverse ' one, by studying the hot ductility of steel on worked products.

Fig. 15 shows this fact by means of the variations in ductility as a function of temperature obtained on tensile test-pieces taken either longitudinally or transversely from a 90 mm. (3½ in.) bloom, annealed, made from a 2-ton ingot of a 5% Cr : 0·5% Mo steel. It will be seen that the steel has a much reduced ductility in the transverse direction than in the longitudinal one.

The method of taking test-pieces has, therefore, a great practical importance in the experimental study of the hot working of steels. It has been already pointed out indeed that, in certain cases, the steel is essentially worked in the transverse direction (particularly in the case of tube rolling mills of the Mannesman or Stiefel type) and it is known in practice that, in order to get at the hot behaviour of steels in inclined tube mills, it is only by studying the transverse ductility of the steel that valid conclusions can be obtained. It should be remarked in this connection that a torsion test subjects the steel to shearing stresses, which in the case of a test-piece taken lengthwise from a worked steel, stresses the metal in a transverse direction. The number of turns before breaking in such a test therefore shows the transverse hot ductility of the alloy steel being studied, and this may explain the interest which tube makers have generally shown in this kind of test.[3–6]

Influence of section reduction by working

Working improves the hot ductility of steel in the longitudinal direction. To provide evidence of this, tensile test-pieces have been taken lengthwise from a 30 kg. (66 lb.) ingot, 100 mm. (4 in.) square and from 2, 1 and ½ in. square bars forged down from similar 30 kg. ingots, all from the same cast of an 18/8 type of austenitic steel. All the test-pieces have been taken tangentially to the surface of the different products studied. In view of the considerable influence of the preliminary treatment on the hot ductility of this steel, all the test-pieces have been subjected to the same preliminary heat treatment, namely air quenched at 1,050°C. for 1 h.

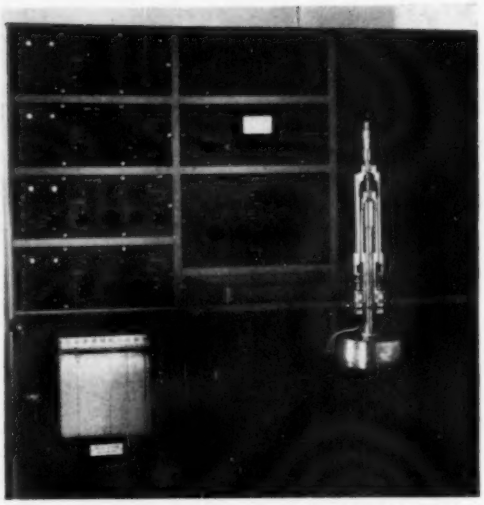
In fig. 16 will be found the curves for hot ductility taken by drawing on the various products with increasing amounts of section reduction. These graphs show that working improves deformability on the length, in proportion as the temperature is higher. It will be noted that deformability of the steel as cast gives two different curves. The tensile test on this steel produces, in fact, a distorted fracture, of which the cross section is approximately an ellipse. The minor axis of this ellipse is parallel to the dendrite crystals. As the test-pieces were taken near the surface of the ingot, the upper curve consequently corresponds to the deformability parallel to the dendritic axis and the lower curve to the deformability in a perpendicular direction. Working effaces this anisotropism.

Different phenomena can be observed if the influence of working on the hot ductility of a steel is studied in the transverse direction. Although some steels show a continuous improvement in their hot ductility, as working increases in degree, both in the transverse and longitudinal directions, it is found in some cases that transverse ductility decreases as working increases, without the longitudinal ductility being affected. This fact is shown in fig. 17 for products with increasing amounts of section reduction, taken from a 2-ton ingot of a mild steel with 5% Cr and 0·5% Mo. This graph shows the variations of ductility with test temperature obtained on tensile test-pieces taken crossways on annealed blooms 190, 140 and 90 mm. square (7½, 5½ and 3½ in. square) (curves a, b and c). These test-pieces were taken parallel to the surfaces of the blooms so that the deformed area is near to the geometrical axis of the various work-pieces studied. Curve L shows the longitudinal ductility taken on a ½ in. square forged bar from the same ingot.

It is thus shown that the same laws for hot transverse ductility of steels exist as for the more usual phenomena already established for cold transverse ductility.[12, 13, 14]

Metals research at the N.P.L.

The primary work of the Metallurgy Division of the National Physical Laboratory, Teddington, is to investigate the constitution, structure, and physical and mechanical properties of both ferrous and non-ferrous metals and alloys, in order to provide information for a better understanding of the factors controlling their behaviour and thus to facilitate the development of new alloys and the most efficient use of existing ones. Methods of investigation range from examination of the structure of a metal or alloy by means of both optical and electron microscopy, and by X-ray and electron diffraction, to mechanical, thermal and electrical tests of the material. The development and application of new experimental techniques for the study of metallurgical problems receive special attention. Some of the work in progress described below was exhibited at this year's open days.



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1. *Equipment for measurement of creep strains*

FOR A NUMBER OF YEARS the Metallurgy Division of the N.P.L. was actively engaged in the study and development of alloys of aluminium, and, later, of magnesium alloys. It was largely due to the Laboratory that the scientific study of light alloys for aircraft production was started some 38 years ago. Investigations of the phenomena taking place during the age-hardening process in these alloys have been an important feature of subsequent research. Alloys of magnesium containing cerium and small proportions of other metals have proved to have exceptional properties at elevated temperatures, and their constitution and creep resistance have been systematically studied.

More recently the possibilities of titanium, a metal somewhat heavier than aluminium but very much stronger and with a considerably higher melting point, have been generally recognized. Work has been undertaken on the chemistry of processes by which it may be extracted, and on the effects of impurities and alloying elements on its structure and properties. Another metal which has specially received attention in the Division is niobium. In this case interest centres not only in niobium metal itself and in alloys in which niobium is the major constituent, but also in the beneficial effects that can be produced by small additions of niobium to alloys for use at high temperatures.

Another important activity is the preparation of metals of high purity, some of which are not very well known industrially. These have included beryllium, chromium, cobalt, iron, magnesium, manganese, silicon and vanadium. The products serve for the determination of the physical properties of the pure metals, and for the study of alloy systems free from impurities.

In recent years emphasis has been on the production of relatively large amounts of pure iron. This was made possible by the installation of a high-frequency electric induction furnace in which it is possible to melt and cast 50 lb. of the metal in a high vacuum. The material is used in investigations into the effect of alloying elements on the properties of iron, including in particular a study of the factors affecting the low-temperature embrittlement of the metal, a phenomenon of considerable practical importance in the operation of engineering structures in cold climates. Much of this work is carried out on single crystals which are produced in the Division. Single crystals of other pure metals and of solid solution alloys are also made for use in a study of the relationship between the elastic constants of the material and its composition.

Mechanical properties testing

The increasing demands made on plant to withstand higher service temperatures and stresses lead to a continuous search for improved or new materials for the construction of power plants whether of the nuclear or more traditional types. Sometimes the demand is for materials to withstand fatigue at high temperatures, and at others

2 ULTRASONIC GONIOMETER FOR TEXTURE MEASUREMENT ON METALS. *The device is intended to examine certain metals, such as nimonic, titanium, brass, copper, uranium, etc., of marked elastic anisotropy which are often fabricated in forms exhibiting appreciable texture, that is, preferred orientation of crystallites; it will assess the extent of this and also its directionality, e.g. the determination of the rolling direction of a plate. It achieves this by observing the critical angle of reflection of an ultrasonic wave incident on an area of about 1 sq. cm. of the surface of the specimen; access to the rest of the specimen is unnecessary, and, apart from the test surface, the shape of the specimen is unimportant provided that it is not too thin. The surface examined can be cylindrical, and, in certain circumstances, curved somewhat in any direction.*

The device itself requires an input electrical wave train to produce the beam and an amplifier to take the received signal from the piezo-electric crystal so that it can be displayed on a cathode ray tube. Both these functions can be performed by a good quality ultrasonic flaw detection equipment.

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for materials of low density for engine casings, etc. The properties of engineering materials at elevated temperatures are studied in the high-temperature mechanical properties section. Two laboratories with controlled ambient temperature are equipped with approximately 90 creep and relaxation machines while another has some 20 fatigue machines of various types.

The creep machines have been designed to measure the slow deformation of metals under steady loads in the temperature range from 20°C. to 1,250°C. for periods up to 100,000 hours, and investigations to provide the designer with such long-time data are in progress. The machines are semi-automatic and can detect creep strains of 10^{-8} per hour. Since the failure of a component in service at high temperature may be due to fatigue of the material under alternating or fluctuating loads, the effects of cyclic loading on the life of engineering materials are being studied over a range of temperature up to 900°C.

In parallel with this general study of the creep and fatigue properties of metals under load at elevated temperatures, attention has been directed to the effects on these properties of different atmospheres such as occur, for example, in steam power plant. This enables materials in the development stage to be compared for general suitability and for freedom from stress-corrosiou effects.

Electrical strain-recording apparatus for creep testing.—An apparatus has been developed to meet the need for a technique providing more efficient means of obtaining strain data from a large number of creep machines than the conventional ' rhomb and mirror ' optical lever high-temperature extensometer. A prototype was shown at the Physical Society Exhibition this year, and following trials with the prototype a pilot scheme is in hand that involves a small number of creep machines used for medium and high sensitivity creep testing. The operation of these machines, equipped with the electrical extensometer, will provide information about the savings in labour used to obtain strain data, the maintenance of extensometer calibrations over long periods of testing and the practical usefulness of the recording facility (Fig. 1).

Creep deformations normally measured range from minimum extensions of approximately 0·00002 in. up to total extensions of 0·20 in. and over. These deformations are detected in a test-piece with a high-temperature extensometer and the movements converted into changes of mutual inductance by a linear differential transformer, in which an iron core moves relative to a primary and two secondary windings. The secondary windings form part of a simple bridge which can be balanced by resistances in the other parts of the bridge and at balance the changes of resistance are a measure of the movements detected. By adjusting the bridge parameters a given change of resistance can be made numerically equivalent to a given test-piece strain, and a direct reading decimal scale of strain may be provided for a variety of test-piece

gauge lengths and sensitivities of measurement. The bridge is energized from a 50 c/s. a.c. supply which facilitates automatic balancing with a slide-wire resistance driven by a conventional servo-motor.

Research into constructional materials cannot be conducted by laboratory experiment alone and the Division is accordingly equipped with a foundry and rolling mill for the production and working of metals on a small scale.

The X-ray and Metal Physics Section is primarily engaged in fundamental researches; it also collaborates in other investigations in the Laboratory when the special techniques of X-ray diffraction or electron microscopy are required.

One of the researches in progress using a soft X-ray spectrometer designed and built in the Laboratory has as its objective a better understanding of the nature of the cohesive forces between the atoms of metals. X-ray studies of the deformation of metals have been carried out over a number of years and recent investigations are concerned with the mechanisms by which preferred orientation of the crystal grains develops as a result of the deformation produced in a metal during such fabrication processes as forging or rolling. The section is actively co-operating with the International Union of Crystallography in its project to determine the precision with which fundamental crystal lattice dimensions can be determined by X-ray methods. New equipment such as high-temperature X-ray cameras, fine-focus X-ray tubes, etc., are developed as necessity for improved apparatus arises. A very recent project is the design and development of an apparatus for the analysis of metals by X-ray emission analysis.

The changes of microstructure and the deformations which take place during the creep of metals and alloys are under investigation. The use of the various types of microscope permits the changes of structure and the subdivision of crystals to be followed and sensitive measurements of slip and grain boundary movements to be made.

Electron microscopy

The recent acquisition of a new electron microscope having a resolution of 10Å (10^{-6} mm.) or better has made considerable advances possible so that, for example, the actual movements of the dislocations in metals under stress can now be followed visually. Interesting observations have been made on alpha iron using a technique which in effect allows dislocation, precipitates and other imperfections to be seen in silhouette; what are believed to be the first detailed photographs of precipitates on dislocations have been taken. The precipitates grow on the dislocations under suitable conditions and develop fern-like or dendritic shapes, although the branches are only about 100 atoms thick. The dislocation arrangement itself is based on a simple structural unit, namely, a threefold junction, but because the units are put together in different ways a variety of dislocation patterns pervades the metal. The observations made so far indicate that the properties of the unit are a major factor controlling the yield and creep strength of iron and also materially influence the behaviour on annealing.

Experiments are also in hand to measure the energy of a metal surface and also that between the grains (grain boundary energy) in a polycrystalline material. Measurements at different temperatures of the rates of diffusion of the atoms within a pure metal (self-diffusion) and of the atoms of one metal in another are in progress, using radioactive tracer techniques to follow the movements of the atoms. At the same time the effect of stress on the rate of diffusion is being studied by a comparison of the movements in two similar specimens held at the same temperature, one stressed and the other unstressed.

The Chemistry Section provides facilities for all analyses required in the Laboratory and attention is paid to the development of new and more sensitive analytical techniques. In particular, effort is now being directed to an assessment of the usefulness of neutron activation analysis. The basis of the method is that when a metal is irradiated with neutrons in an atomic reactor radioactive isotopes of the various impurities are produced. Each decays at its own characteristic rate with the emission of β or γ rays of characteristic energy, so that with suitable electronic counting equipment analysis of the sample is made possible. This method is particularly useful in the analysis, for example, of semi-conductors such as silicon and germanium in which impurity levels of 1 in 10^9 or 10^{10} have marked effects on the electrical properties. Traditional methods of chemical analysis are quite inadequate for materials of this order of purity.

A comparatively recent widening of the field of interest is provided by a small group working on the thermodynamics of metals and alloys. Measurements of vapour pressures, heats of formation, etc., are under way with the object of accumulating sufficient data to enable, for example, reasonably accurate forecasts of the equilibrium constitutions of alloy systems to be made, particularly at low temperatures where the traditional methods of thermal analysis fail because the attainment of equilibrium is so slow.

Special investigations are undertaken for individual firms into problems arising in their manufacturing or engineering processes, and examinations are made of the causes of failure in service of structural components.

Production of slugs and forging blanks

Automatic 'displacement' diecasting method

THE CONVENTIONAL METHOD of producing slugs for impact extrusion is to cut them from rod stock, and blanks for forgings are also, for the most part, sawn or sheared from rod or bar. Since in both cases the original shape of the stock is completely lost in the forming operation, the cost of bringing it to that shape represents a proportion of total production costs which is only justified if there is no simpler and cheaper way of producing a blank. Additionally, in the production of small slugs and forging billets there is a considerable wastage of metal due to the width of the saw-cut, short ends and so on.

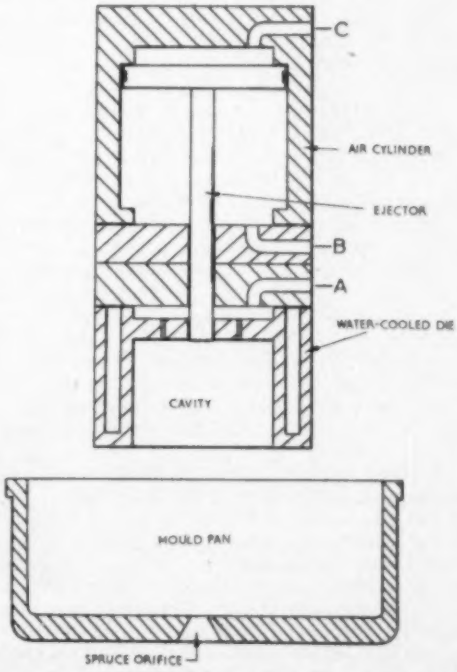
Although various foundry processes offer obvious alternatives for the production of blanks, the internal structure of cast parts is often unsuited to further working, whilst the surface often carries abrasive matter deleterious to the forging dies. As the size of the blank diminishes, the possibility of using cast blanks vanishes, since with increasing surface/volume ratio the chilling rate is faster and the feeding in of extra metal to compensate for shrinkage becomes, with conventional casting methods, more and more difficult.

During the last few years there has been a very rapid growth of both impact extrusion—especially of aluminium—and of small forging production in aluminium alloys. This has raised with some acuteness the question of economic slug production, and the Stroman Furnace and Engineering Co., of Franklin Park, Ill., has recently introduced equipment capable of producing automatically individual slugs and blanks in a wide variety of shapes to very close weight limits. The machine is self-contained; ingot or scrap is fed in at one end and the slugs are delivered at the other.

'Displacement' diecasting

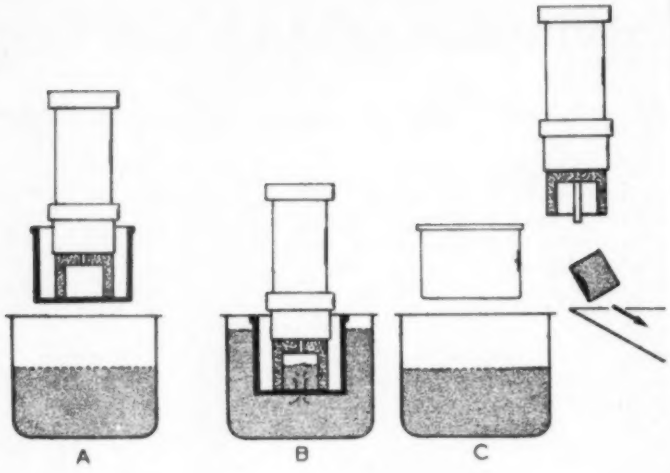
The process is in many respects a novel one, and the 'Como-Cast' automatic slug-casting machine is, in effect, a diecasting machine operating on the 'displacement' principle. The entry of metal into the die cavity, that is, is effected neither by air-pressure nor suction, but by the hydrostatic head of the molten metal. Because metal is taken from well below the surface of the melt, there is little chance of dross being carried into the cavity. By an ingenious refinement of design, the diecast slugs are sheared from their sprues before ejection and are thus ready for use without trimming when delivered from the machine.

The 'Como-Cast' machine consists essentially of a melting furnace feeding a holding pot; above the latter is mounted a light machine-frame of conventional tie-bar design and carrying a single, moving, die platen. The axis of die travel is vertical, as shown in fig. 1. This whole assembly can be moved laterally to a position clear of the holding pot. The die, which for cylindrical slugs is of simple pot form, is water-cooled and is mounted on the face of a short-stroke pneumatic ejector cylinder. The lower end of the cavity is open. Below the die, and independently actuated, is the 'mould pan' which serves to close the die cavity. In line with the cavity is a sprue orifice in the pan bottom; this has reverse taper as shown. The height of the mould pan wall is such that at closure the top face of the cavity is below the rim of the pan.



1 *Stroman 'Como-Cast' machine for automatic slug production. Air displaced by the incoming metal escapes at* A*; inlets* B *and* C *serve the pneumatic ejector*

2 *Operating sequence of the slug caster*



In operation, the die member is brought down into the pan and presses tightly upon the bottom (fig. 2 A). The whole assembly is then lowered steadily into the holding pot, displacing metal and so raising the level in the pot. At the same time metal rises through the sprue-hole and begins to fill the die cavity (fig. 2 B), the air escaping by venting channels at the top of the pot ahead of the rising metal. The metal chills as it contacts the top face of the cavity and so does not plug the vents. If faster filling is required it is possible to couple a vacuum line to the outlet at *A*, fig. 1.

Because filling is from the bottom, whilst solidification proceeds inward and downward, crystallization shrinkage can be completely compensated and, despite the relatively low pressure at which they are diecast, the slugs are dense and homogeneous. The rate of solidification is roughly 12 to 14 sec. per in. of slug diameter, though this, of course, depends to some extent on slug thickness. Once the correct dwell has been determined, a pre-set timer ensures that it is correctly repeated at each cycle.

As soon as solidification is complete, the die assembly is lifted clear of the melt again. The mould pan is now locked in place, and the die is shifted sideways a small distance while still in contact with the pan bottom. This shears the small sprue remnant in the upper part of the orifice and it drops back into the pot. The die is now lifted right out of the mould pan and the die carriage is traversed sideways clear of the pot. The pneumatic ejectors then thrust the slug from the cavity (fig. 2 C), whence it falls into a discharge chute. The ejectors retract, the die carriage moves over into alignment with the mould pan, and a new cycle commences. The holding pot is replenished from the melting hearth as required.

Advantages of the method

By this process of slug casting there is virtually no waste at all, for the trimmed scrap from the forming process can be returned for immediate re-melting. Although for clarity only a single impression die unit is shown in the figures, in practice it is highly advantageous to adopt multi-cavity dies. Each cavity, of course, is fed from an independent sprue orifice in the mould pan. As many as 16 slugs per cycle have been produced successfully with the equipment. Although the cycle is slow as compared with that of most impact extrusion operations, therefore, the output can often match the latter and the equipment may accordingly take its place in an automated production line. Most impact extrusions are worked from a disc-like slug, but for forgings, where such extreme deformation is not desirable, the blank may advantageously be diecast in a form approximating to that of the finished part. Except that undercuts cannot be produced and that adequate draft for ejection is provided, dies are not restricted as to form. It is, therefore, quite practicable to diecast a 'use' so contoured as to be worked evenly all over under the forging die and to come to finished shape with a minimum of flash. Although the primary demand for the machines comes from producers of light alloy extrusions and forgings, other non-ferrous metals can be successfully handled without modification apart from re-timing to suit thermal variables.

Embrittlement of solid metals in a liquid metal

W. A. MORGAN

This review of the published work on the embrittlement of solid metals by lower melting-point liquid metals is published by permission of the Director, Mines Branch, Department of Mines and Technical Surveys, Ottawa, Ontario, Canada. The author is Head, Ferrous Metals Section, Physical Metallurgy Division of the Department.

MANY SERVICE FAILURES resulting from soldering and brazing operations on stressed materials have been reported. Generally the fracture surfaces followed an intercrystalline path and were often wetted by the lower melting-point metal. This review is a summary of the published work on the embrittlement of solid metals by lower melting-point liquid metals and discusses some of the theories for this type of failure.

The intergranular failure of high melting-point metals when stressed in contact with a low melting-point molten metal or alloy has often been compared with the phenomenon of season-cracking. Possibly the similarity is borne out by the fact that Rogers[1] developed the use of mercury-salt solutions for the detection of internal stresses in brass. The simultaneous effects of internal stresses in the material and the interaction of deposited mercury with the grain boundaries of the brass causes cracking along intercrystalline paths. The method was used by Heyn[2] and later thoroughly investigated by Rawdon.[3] Huntingdon[4] reported the complete intercrystalline disintegration of a β-brass containing about 2% aluminium when treated with mercury. Desch[5] obtained individual crystals from a similar brass by immersing it in mercury, but was unable to obtain the same effect with zinc, tin or lead. Moore and Beckinsale[6] showed that brass could be made immune from mercurial penetration and season-cracking by a low-temperature anneal.

Mercury penetration

A very thorough investigation into the whole problem of season cracking was reported by Moore, Beckinsale and Mallinson[7] in 1921. The authors quoted results of mechanical tests on specimens coated with mercury which had been deposited from a mercurial salt. They found that the mercury alloys with the brass, but the action was confined to only a very thin surface layer. Fracture began at grain boundaries on the surface of the specimens and followed the crystal surfaces which became covered with mercury. The level of stress at which the intercrystalline cracks formed was shown to be dependent on the amount of free mercury and the time available for penetration. Slower rates of loading give lower values for the maximum stress and elongation. Very little deleterious effect was brought about by immersion in mercury and then testing after removal of the liquid. They concluded that little or no intergranular penetration occurred in the absence of stress. Finally the results of tests are given which show that copper was not embrittled by mercury, but that the susceptibility to failure increased with the zinc content.

Edmunds[8] confirmed this finding in the range 10—40% zinc. In his investigation he also showed that 'as-rolled' mercury-coated tensile specimens and specimens of large section possessed superior resistance to penetration than recrystallized and small-section test-pieces. He correlated the grain size of 70/30 brass with its mechanical properties in mercury, showing that the properties decreased with increase of grain size. A single crystal, however, failed in a completely ductile manner and showed no deleterious effect.

Investigations into the effect of alloying elements on the susceptibility to penetration of brass by mercury have shown that no really marked effect was produced. Lynes[9] tested annealed 70/30 brass containing 0·1 and 0·3% arsenic, and Bassett[10] investigated the effect of 0—0·2% cadmium. Wilson, Edmunds, Anderson and Peirce[11] alloyed 36 elements with cartridge brass. In all cases the authors concluded that none of the added elements accelerated failure and the beneficial effect, if any, was small.

Strachan, Jones and Harris[52] studied the effect of prior immersion in mercury at 300°C. and 500°C. on the room temperature mechanical properties of Inconel (80 Ni, 14 Cr, 6 Fe), Nimonic 75 (75 Ni, 20 Cr, 5 Fe), nickel-iron 50 Ni/50 Fe, an 18/8 stainless steel, a 3·5 Si/8 Cr valve steel, mild steel sheet, a pure iron sheet produced from powder, molybdenum and tungsten. At 300°C. no changes

in the mechanical properties were observed. At 500°C., however, nickel was the only material which showed a decrease in mechanical properties. Further tests on the 20 Ni/25 Cr steel and the mild steel, after immersion in mercury at 300°C. for 2,000h. under a stress of 12,200 lb./sq. in., showed no decrease in mechanical properties. An interesting facet of this work was the effect of 30 kc./s. ultrasonic vibrations on these materials. The authors showed that they were all susceptible to cavitation-erosion in mercury at room temperature.

Penetration of molten solders

The researches of the early workers on mercury penetration into brass and season-cracking led Dickenson[12] to report the failure of a manganese bronze by intercrystalline penetration of molten solder. He later published the first account[13] of an investigation into the embrittlement of a solid metal by a low melting-point metal. The results of his tests showed that the behaviour of different brasses varied with their microstructure. With a tin/lead solder a β-brass failed at a lower stress than an α-brass, an α + β brass occupying a midway position. For molten lead/tin/bismuth alloys containing 50% bismuth, α-brasses were much more seriously affected. Dickenson[13] offered no explanation for this. Further examples of failure as a result of such penetration followed; the embrittlement of malleable cast iron by zinc in hot-dip galvanizing was reported.[14] Duncan[15] found that molten brass penetrated nickel steel. Hall[16] reported the embrittlement of a mild steel by solder under alternating stresses at a temperature below 60°C. It is surprising that the penetration should have occurred at a temperature when the solder is solid. Probably some overheating and localized melting must have occurred at the point of penetration.

In 1927, Miller,[17], Hartley[18] and Genders[19] published three papers covering a wide range of nonferrous and ferrous metals. Miller tested specimens of 60/40 and 70/30 brasses in mercury at room temperature and in 50% lead—50% tin solder, pure tin and pure lead at temperatures of 220°C. and 350°C. He found that failure occurred by a process of intercrystalline penetration, and concluded that although a much higher stress was necessary for brass to crack in molten tin solder or lead, the process was similar to that of cracking and embrittlement by mercury. In agreement with previous workers his results showed that α-brass was more resistant than β-brass to penetration by mercury and lead-tin solder.

Hartley's work was similar to that of Miller except that he extended his range of copper/zinc alloys from pure copper to 60/40 brass. In addition, he measured the time to fracture at various stresses and compared the results obtained with values from short-time tensile tests in air at the same temperature. An analysis of his results showed that the strength of pure copper was not so drastically reduced in molten tin as were the 70/30, 64/36 and 61/39 brasses, the exception being the 80/20 brass. The stress required to fracture a test-piece was time-dependent and the stress-duration curves were of parabolic form. The author showed that the tin content of the solders used had an effect on the penetration, increasing tin content giving more rapid failure. On this finding, he based an explanation of the failure by postulating that alloying occurred at some grain boundaries which contained segregated impurities more rapidly than at others. In this way a brittle film may be formed which fractured easily under the applied stress; at this point further alloying and subsequent fracture took place.

Effect of brazing solders

Genders studied the failure of mild steel when coated with commercial brazing solder. A ¼-in. rod could be fractured easily by bending it at 800°C. Molten copper also produced such an effect, but surprisingly he reported that zinc, tin and tin/lead alloy had no effect. He indicated that the experiments he had carried out showed a small amount of penetration of brazing solder at grain boundaries of armco iron in the absence of stress. In a discussion on Genders' paper, Williams[20] reported the failure of a 50/50 nickel-iron alloy in brazing solder. Anderson and Poole[21] quoted results showing that molten copper was much more severe than brazing metal on mild steel and that, although zinc had the strongest solvent action on steel, it did not give rise to embrittlement. They showed that a critical stress level had to be exceeded to produce embrittlement.

Riede[22] described cracks which developed in steel during dip-brazing and showed that normalizing could be used as a preventative. Stresses which arise during the hot-dipping process have been suggested by Albert[23] as a cause of penetration. Kreitz[24] found cracks after brazing articles which he believed to be free from internal and external stresses. Failures due to intercrystalline penetration have been reported by Ffield[25] in the welding of certain alloy steels which had been galvanized. Sparagen and Claussen[26] extended these findings to include both plain and alloy steels. Straight chromium and 18/8 stainless steels have been shown by Henry and Schroeder[27] to be susceptible to embrittlement by zinc, tin and Tobin bronze in both the molten and gaseous states. Zinc showed the most severe attack. The use of zinc coatings on nickel/chromium steels was shown to be undesirable if they were to be subsequently heated.[28] In order to determine at what temperature intergranular

penetration of galvanized steel by zinc might take place, Craig[29] conducted a series of high-temperature tests on coated and uncoated steels. He found that in the temperature range from 20—800°C. low-carbon and high-carbon steels suffered no adverse effect due to zinc coatings. In stainless steel the coating had no effect up to 550°C. Above this temperature the zinc coating appeared to cause a marked decrease in ductility, but a less pronounced loss in strength.

Effect of other metals

Schottky, Schichtel and Stolle[30] investigated the effect of a whole range of metals on plain carbon steels, 4% silicon steel and on 22% and 32% chromium steels. The tests were made by bending bars, in the range 1,000—1,200°C. and sprinkling metal powder on the side in tension. Tin, zinc, antimony, copper, 5% tin bronze, a 5% tin, 10% zinc brass, and white bearing metals caused marked red-shortness at all temperatures in the above range. Bismuth, cadmium, lead, silver, and nickel seemed to have little or no effect. The carbon content in the range 0·1—0·5% carbon had no effect on the resistance to penetration. No red-shortness was observed in the 32% chromium steel with any of the metals used. A short time after this investigation was published, Schuster[31] reported the penetration of brass into a steel shaft.

Van Ewijk[32] in 1935 reported that stressed nickel-chromium high-tensile steels were embrittled by molten solder, lead, tin, zinc, cadmium or Tainton alloy (tin 4, lead 8, bismuth 15, cadmium 8 parts; melting point 60°C.). For carbon steels no such liability was observed. In order to obtain wetting of the steels he used a zinc chloride flux; no cracking was observed on specimens in which the flux had not been used. The liability to failure in the molten metal was greatly reduced by tempering at 700°C., but normal ranges of tempering temperature produced little effect. Pattermann[33] confirmed Van Ewijk's[32] results and reported that chromium/molybdenum steel was much more resistant to embrittlement than the nickel/chromium and nickel/chromium/molybdenum steels. Rustproof chromium steels and nickel/chromium/titanium steels only showed the effect to a very slight degree.

Variables affecting penetration

An investigation of such variables as temperature, rate of application of load, tensile properties, degree of temper-brittleness and microstructure has been made for the attack of molten white metal on steels, by Goodrich.[34] By means of a bend test he plotted load deflection curves for four different steels uncoated and coated with lead/tin solder at 250°C. and a bearing metal at 350°C. The steels used were an oil-hardened and tempered plain carbon, 1% nickel and nickel/chromium/molybdenum steel. With increase in the rate of loading for these materials there was a gradual decrease in the yield and maximum loads, the smallest decrease being shown by the plain carbon steel. An increase of temperature brought about a more rapid decrease of these properties and again the plain carbon steel seemed more resistant to penetration.

Goodrich further showed that addition of nickel up to 1·2% to carbon steels was deleterious. However, the resistance of nickel-chromium steels was either better or worse than carbon steels depending on the microstructure, a small amount of free ferrite and small sorbitic grains decreasing the liability to embrittlement. Generally, carbon steels were more uniformly resistant to penetration than alloy steels. Quenched and tempered structures gave better results than normalized or annealed structures. Addition of molybdenum to the nickel/chromium steels conferred no added resistance. In agreement with other workers, Goodrich reported that the resistance to penetration varied inversely as the grain size. Finally, temper brittleness, as determined by the Izod impact test, appeared to have no influence on the resistance to penetration.

The resistance to penetration of steels by molten white bearing-metal solder and brazing alloys was also investigated by Wang.[35] His results confirmed that alloy steels were more susceptible than plain carbon steels. Intercrystalline attack by low melting-point metals did not occur above about 400°C. The test of the load extension curves obtained showed that penetration did not occur until the elastic limit had been exceeded, and that the degree of stress required to produce penetration was critical. The rate of application of load when increased, increased the tendency of molten brazing alloys to penetrate steels; the tendency for such penetration was diminished above 950°C.

Comparative tests in solder and oil

A very extensive range of engineering materials were tested in molten solder and oil by Austin.[36] His tests were carried out in a tensile testing machine and a complete autographic load-extension curve was plotted for each specimen. In this way it was possible to determine at what stage of the curve (elastic, general extension to maximum load, or necking) the effect, if any, occurred and whether the normal tensile flow curve was interrupted only, or modified. Generally, it was shown that the molten metal did not modify the curve, but stopped it earlier than comparative tests in oil. The presence of molten solder, therefore, did not apparently affect the general resistance of the material to deformation, but decreased the amount of deformation before fracture.

Iron, nickel and copper were slightly affected; monel metal and cupro-nickel were unaffected. The strength of silver, zinc, aluminium, aluminium/bronze, 70/30 and 60/40 brass was reduced. The plain, low-carbon steels and the lower carbon pearlitic and austenitic heat- and rust-resisting steels did not appear to be as liable to penetration as similar steels with higher carbon. Increasing hardness and secondary grain size were deleterious, but neither the temper brittle nor the burnt state appeared to increase the embrittlement. Bassett[37], in the discussion on this paper, reported the penetration of bearing metal into a steel railway wagon axle due to local overheating and alternating stress.

The most recent report on the effect of molten solders on stressed solid metals has been published by Russian investigators.[54, 55] They studied the influence of composition and contact time of various solders on the brittle fracture of steels in tension. A beneficial effect was claimed for surface protection by nickel and copper. In the case of copper this differed from the results of Jepson and Thompson[38] discussed later in this report.

Effect of surface finish

Only a very small amount of information existed on the effect of surface finish of steel on its liability to penetration. Most of the above workers found that unless they obtained wetting of the steel by the solder with a zinc chloride, flux failure was of the normal ductile type. However, Jepson and Thompson,[38] while investigating the effect of a tensile stress on the rate of transformation of a eutectoid steel under isothermal conditions, found that specimens immersed in molten solder at temperatures between 200 and 400°C. were liable to fracture under low stresses. They, therefore, determined the time to fracture under a given load at 250°C. of specimens, each having a different surface finish. Their results showed that nickel-plating was the only effective prevention to failure, but even this failed at high stresses. An oxide skin proved the next safest and a copper-plated specimen the worst. When lead was used instead of solder no embrittlement of the steel occurred.

In the field of galvanizing particular attention has been paid to the interface reactions between molten zinc, the steel being dipped and the molten zinc bath-container materials. In 1952 Haarman[50] reported that galvanizing baths may be subject to three types of failure due to the attack of molten zinc. They were general surface attack, deep pitting and intercrystalline embrittlement. Initial results of his work showed that the higher the stress the greater was the degree of embrittlement.

Later, a more extensive investigation was reported by Radeker[51] who again showed that steel is subject to intercrystalline attack by molten zinc. Bent specimens of steel immersed in zinc baths at various temperatures between 425 and 500°C., with intermittent closing of the U-shaped strip, revealed no cracks until a notch was introduced at the bend, thus demonstrating the need for a minimum stress level. Both tensile strengths and, particularly, the reduction of area were decreased as a result of intercrystalline attack at temperatures of 425, 450, 475 and 500°C. At higher temperatures and stresses, stress-rupture tests on specimens immersed in molten zinc showed that the life was shorter than at lower stresses and temperature. The relationship between stress and time yielded a straight line on a double-log plot.

The results published by Hoarman[50] and Radeker[51] differ from earlier results of Anderson and Poole who reported no intercrystalline embrittlement due to zinc. However, the former investigators agree with the work of Ffield, Henry and Schroeder, Campbell, and Craig.

Penetration of non-ferrous alloys

Service failures in four groups of non-ferrous alloys have been reported by Smith and Forsyth.[39] A high-strength brass containing aluminium having a homogeneous β-structure was easily penetrated by solder, whilst a similar brass containing manganese and possessing a duplex α-β structure was unaffected. Cracking was also observed in a completely β-brass containing lead. During annealing the lead had formed thin films around the grain boundaries and had ceased to exist as discrete globules. This type of failure has been observed to occur during the working at elevated temperatures of leaded steels.[40] Fracture of high-strength nickel-chromium steel tubes by penetration of molten solder and of austenitic steel by brazing alloy at 300°C. was also reported by Smith and Forsyth. They showed, too, that an aluminium/copper/magnesium/zinc alloy may be embrittled by Wood's metal, and that platinum alloy electrodes were susceptible to penetration by lead deposited on them during operation.

The Aluminium Development Association[41] investigated the embrittlement of aluminium and aluminium alloys in solder and brazing metal (9% tin, 10% zinc) at 250°C. Pure aluminium and aluminium/manganese alloy or annealed aluminium/magnesium/silicon alloy did not fail intergranularly. Fully heat-treated aluminium/magnesium/silicon alloy and a fully heat-treated aluminium/magnesium/zinc alloy were found to be the most susceptible.

Whitaker[42] reported the work of Voce and Bailey on the embrittlement of α-, α-β- and β-brasses up to 4% aluminium in mercury, lead, tin, and eutectic solder. With mercury, fracture occurred in all the brasses, the β-brass being most susceptible. The

stress at failure increased with the aluminium content. At 200°C. each brass was embrittled by solder and again the β-brass was the most susceptible. However, in the α-β alloys preferential alloying was reported to have occurred between the α phase and the solder. Contrary to this finding, at 350°C., the alloying occurred equally with the α and β phases. Increase in the aluminium content conferred no added resistance to penetration. Lead at 350°C. did not penetrate plain 60/40, 70/30 and 53/47 brasses.

Stainless steels

Although many investigations sponsored by the U.S. Atomic Energy Commission on the solution of solid metals in molten metals have been published recently, only a comparatively small number have dealt with stressed metals. Grassi, Bainbridge and Elliot[45] investigated the stress-rupture properties of five low-alloy and stainless steels in liquid lead-bismuth alloy at 750°C. When the metals were not pre-coated before immersion in the molten metal no significant attack was noted. However, coating before testing resulted in a small decrease of the properties of the steels. The greatest decrease in strength was shown by a chromium/molybdenum/vanadium steel.

In a later publication the same authors[46] tested niobium and molybdenum in lead and bismuth. They reported that an anomalous increase in strength was noted for molybdenum in the stress-rupture tests at 850 and 750°C. in the molten metals. No significant surface attack by the liquid metals was noted. Niobium on the other hand showed a decreased strength and ductility, bismuth having the greater effect and producing fissures in the surface of the metal. The welding of niobium to molybdenum resulted in a zone between the two metals, which showed the development of large grains and broad grain boundaries containing some unknown precipitate. These boundaries were particularly susceptible to embrittlement by molten bismuth.

Pray, Peoples and Boyd[47] reported that stainless steels are relatively unattacked by pure bismuth. Stellite and some of the Haste-alloys are embrittled. Their results indicated that the presence of gases in solutions in the bismuth did not affect the attack on the steels, ' as-received ' bismuth being as reactive as degassed metal. Fatigue and stress-rupture tests on stainless steels in molten sodium at 500°C. have been shown by Koenig and Vandenberg[48] to give results comparable with those obtained in helium and argon.

Penetration by sodium, gallium, potassium and lithium

Sodium, gallium, potassium and lithium have received particular study because of their potentiality as coolants in nuclear reaction. Wilkinson and Yaggee[44] showed that even in unstressed tests Monel, Inconel, mild steel (SAE 1020), Armco iron, nickel types 302, 347 and 440 stainless steels were subject to intergranular attack by lithium at 600°C. No intergranular attack was noted at 300°C. Bokros[56] showed that sodium had no effect on the hot-tensile properties of zirconium. Martin and Smith[57] have also shown that sodium did not affect the fatigue life of 18/8 stainless steels when tested at 300°C. in push-pull tension. On the other hand, these same authors reported that the fatigue limit and the life of mild steel, at stresses above the fatigue limit, was reduced, when tested in molten tin at 300°C.

Herold *et al.*[53] conducted a systematic investigation of the penetration of lithium into pure iron and mild steels (0·1% C.) at temperatures up to 1,100°C. They found that pure iron was resistant to penetration but that steels were embrittled. It was theorized that the reaction between impurities and carbon in steel produced fissures which aided penetration. Heating the steel in hydrogen for 100 h. at 900°C. rendered it impermeable.

The main work on the problem of intercrystalline failure has not been directed to an evaluation of its cause, but rather to seek systems which did not exhibit it. It was not until 1948, when C. S. Smith[49] published an account of the effect of interfacial energies on the equilibrium form of a microstructure, that any further work was carried out.

Effect of interfacial energies

Smith suggested that the interfacial energies between crystals could be considered as tensions acting along the boundaries. When several crystals meet, then their boundaries make definite angles with one another at their point of intersection so that the interfacial tensions formed a balanced system. The angle between two interphase boundaries when a crystal of one phase meets two crystals of another phase was called the dihedral angle. The dihedral angle would vary with the values of the interfacial tensions. The concept could be applied to the interface between a solid surface, solid grain boundary and a molten phase. Smith postulated that if the interphase energy was less than half that of the crystal-boundary energy in one phase, the dihedral angle would be zero and the second phase would tend to spread indefinitely along the crystal boundary.

Furthermore, he concluded that if the dihedral angle was less than 60° a liquid second phase would be expected to penetrate along grain edges (a grain edge being defined as the line forming the junction between three adjacent crystals). He tried to explain the embrittlement of copper-bismuth alloys by this

concept, that since the dihedral angle of bismuth on copper was equal to zero, then molten bismuth would penetrate between the crystals completely wetting the grain faces and disintegrating the solid. Lead would not have the same deleterious effect, since the dihedral angle on copper was much greater than zero. Smith believed that no stress was necessary for such an action to occur. No indication was given of the rate at which liquid bismuth might be expected to penetrate along the grains of a piece of solid copper with which it was in contact.

Robertson[43] has endeavoured to explain the intercrystalline penetration of mercury into brass by applying Smith's ideas. He pointed out that pure copper appeared to be immune from penetration at 170°C. (the temperature above which intermetallic compounds were not stable) since the measured grain boundary groove angle (about 120°) was high. Tests on specimens which were kept under a constant stress while immersed in mercury under the same conditions indicated that the dihedral angle was not appreciably affected by the presence of a stress. An unstressed brass containing 69% copper suffered deep penetration after immersion in mercury for 18 h. at 350°C.; this temperature was chosen since it was above that at which intermetallic compounds were formed. It was suggested that the attainment of the equilibrium dihedral angle at room temperature was prevented by the formation of an intermetallic compound, but that this compound may be broken down under stress.

Discussion

In the foregoing survey details have been given of materials which are susceptible to embrittlement when immersed or coated with a liquid metal. It will also aid in an appreciation of this problem if a brief account is given of the general factors which can effect embrittlement as follows: (1) The surface condition of the solid metal; (2) the grain size of the solid metal; (3) the level of stress; (4) the strain rate; (5) the composition of the molten metal; (6) the effect of temperature.

(1) *Surface condition* One of the prime requisites for embrittlement noted in all publications was that the liquid metal should wet the solid metal surface. Thus oxide films which are not soluble in the liquid metal and which will not be reduced by the liquid metal are protective. At very high deformations, however, the oxide skin may crack, thus presenting fresh unoxidized surfaces to the molten metal. Plating can be protective if the plated metal is not attacked by the molten metal. If it is attacked, then failure is enhanced by the plating, *e.g.* Jepson and Thompson's work. Any method of surface preparation which presents a clean, solid metal surface to the liquid metal can initiate failure in a suitable system.

(2) *Grain size* A large grain size is deleterious. Goodrich showed that the resistance to penetration at 250 and 350°C. varied inversely as the grain size. This effect was very pronounced at 250°C. but not so noticeable at 350°C. Van Ewijk, in a discussion on Goodrich's paper, reported also that embrittlement was increased with a coarsening of the grain size. The confirmation of this effect was provided by Genders, who showed that the degree to which a metal had been cold-worked and the shape of the grain boundaries influenced intercrystalline embrittlement by a liquid metal. The second factor was particularly important at lower temperatures. Desch supported the view that grain shape was more important than grain size. He suggested that intercrystalline embrittlement will not be found in cast metals where there are interlocked and irregular boundaries. This statement, however, seems surprising in the light of Dickenson's results which showed that an as-cast structure of intermingled α and β crystals in a manganese bronze specimen suffered intergranular failure in molten solder. It seems quite conclusive that a large grain-size is deleterious and this may be explained by the fact that crack propagation is easier if the direction of the boundary does not change within short distances.

(3) *Stress level* Many investigations have reported that the time to fracture of a solid metal in a molten metal environment is decreased with increasing stress more so than if the metal was tested in air. Hartley showed that a hyperbolic relationship existed between the stress and time to rupture. Several of the published papers have also indicated that the rate of embrittlement is slow below a critical stress. It appears that this critical stress level may correspond with the yield strength of the solid metal. It is, of course, difficult to determine accurately because local yielding will undoubtedly occur in most metals near their conventional yield or proof stress.

(4) *Strain rate* Goodrich has reported that the yield and maximum loads withstood by different steels in the presence of molten white metal decreased with slower rates of loading at constant temperature. The greatest effect, however, was obtained on the extent of deformation before fracture, which was greatly reduced at lower strain rates. Hartley, too, was able to show that the apparent strength of a solid metal tested in a liquid metal was very much higher in a rapid tensile test than in a slow test extending over about 10 h.

(5) *Composition of the molten metal* Earlier in this discussion it has been pointed out that one of the major factors initiating embrittlement was that the liquid metal should wet the solid metal surface. Lead, for example, does not wet 60/40 and 70/30 brass but solder does. Thus Whitaker showed that these brasses were embrittled by solder at 200°C.

but that lead at 350°C. had no effect. In this instance the tin in the solder promoted wetting of the surface. The wettability of a molten metal is related to the interfacial tension at the liquid/solid interface. If tin wets the surface of solid brass then the lead should readily wet the tinned surface, since tin and lead are mutually soluble.

Other investigators have confirmed this effect on steels where lead produced no intergranular failure at 250°C. but solder did.

(6) *The effect of temperature* Dickenson showed that the maximum strength and deflection of a solid metal tested in bending when coated with a molten metal was decreased by an increase in temperature. The resistance to penetration of steels has been shown to be less at 350°C. than at 250°C. At the higher temperature failure was almost instantaneous, whilst at 250°C., for the same stress the time to fracture was prolonged. Van Ewijk, however, working on the penetration of molten solders into stressed nickel/chromium steels, found that the temperature of the molten metal did not affect the degree of embrittlement. His work was carried out on internally stressed components, so the higher temperatures may have relieved the stress.

One of the more interesting effects of temperature, revealed by Goodrich and Genders, was that the effect of grain size, in lowering the properties of the solid metal in a liquid metal, was greater at low temperatures than at high temperatures. The inference that can be drawn from this behaviour is that the applied stress can decrease the activation energy for grain boundary diffusion and that embrittlement is due to rapid grain boundary diffusion under the influence of stress.

A satisfactory hypothesis for this type of embrittlement has not yet been developed. Some authors believe failure is due to discontinuous cracking and attack along the grain boundaries. However, this idea does not explain all facets of the problem.

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NEWS

CIBA Fellowships

The CIBA Fellowship Trust has been founded for the purpose of improving and increasing the interchange of ideas between scientists in the United Kingdom and on the Continent. The trustees are: Sir Arthur Vere Harvey, C.B.E., M.P., Lord Hives, C.H., Professor Sir Alexander R. Todd, F.R.S.

Several fellowships will be awarded for tenure during the academic year 1959-60 at continental universities or institutions for research in chemistry, physics or some other allied scientific subject. They will be awarded to graduates of universities situated in the United Kingdom of Great Britain and Northern Ireland or in the Republic of Ireland, or to members of those universities graduating this year. It is anticipated that some fellowships will be awarded to recent graduates and other fellowships to candidates who have already taken their Ph.D., or who have already spent some time in industry.

The basic award for Fellows who have already undergone training in research and wish to continue post-doctorate studies will be £800 per annum plus allowances, including cost-of-living and children's allowance. For candidates who have obtained a first degree in science and wish to undergo training in research, the basic award will be £500 per annum plus allowances, including cost-of-living and children's allowance.

Further details are available from the Secretary of the CIBA Fellowship Trust, CIBA (A.R.L.) Ltd., Duxford, Cambridge.

Heat-treatment service

Expert Heat Treatments Ltd. has now established the following divisions for heat treating:

A heavy shop which can take several furnace loads of up to 5 tons and with controlled temperatures of over 1,250°C.

A tool and high-speed steel section together with Martempering facilities having sufficient capacity to handle all normal size tools and dies.

A light alloy section consisting of four forced air circulation, close temperature control, electric furnaces for standard normalizing solutionizing and annealing work, together with a large mechanically charged horizontal furnace for precipitation treatment.

A controlled-atmosphere furnace section in which is installed the latest Efco carbo-nitriding plant and Wild-Barfield gas-carburizing equipment.

Where surface hardnesses are required from direct contact with case-hardening media a cyanide section capable of treating approximately 7,500 lb. of steel per week, and three pack carburizing furnaces with floor areas ranging from 15-28 sq. ft., are in operation.

A tempering section for use where work requires temperature control to fine limits consists of four pot-type forced air circulation electric furnaces.

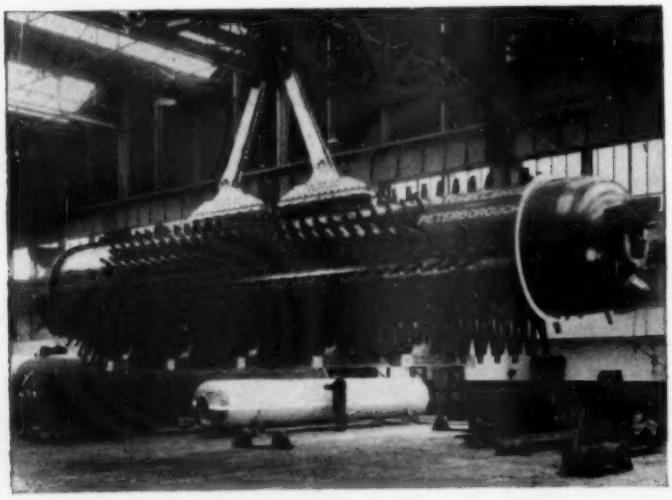
The flame-hardening department has been fitted out to do the most comprehensive range of items possible and is continuously engaged on work varying from hardening the teeth of gears from ½ in. dia. to 5 ft. dia., as well as shafts of all diameters and lengths up to 11 ft. 6 in. long. For this latter purpose special ring burners are in use on a machine which spins the work and traverses the burner at the same time at a pre-determined rate.

The induction department at Morden has a number of machines ranging from 2 kW. to 80 kW. For production runs special fixtures built within the works are utilized and the set-ups are fully automatic with timers and micro-switch controls eliminating any possibility of human error.

British Insulated Callender's Cables Ltd. announces that the address of its Cambridge Branch office is now 65 Devonshire Road, Cambridge.

Chas. H. Windschuegl Ltd., 1 Leadenhall Street, London, E.C.3, has been appointed sole agents in the United Kingdom for 'Beverol' motor and industrial oils and greases manufactured in Holland by Beverol-fabrieken N.V. of Beverwijk, Holland.

The London office of **Rocol Ltd.,** is now at General Buildings, Aldwych, London, W.C.2 (Holborn 1985).



Large forged steel steam drums

First of four forged steel steam drums of an unusual design, produced by Thos. Firth & John Brown Ltd., to the order of Mitchell Engineering Ltd., has recently been installed at the Central Electricity Generating Board's generating station, Skelton Grange, near Leeds. The 85-ton drum, 47 ft. long, has a 66-in. bore, is fitted with 879 welded stub tubes and attachments instead of the usual expanded boiler tubes, varying from 9 to 27 in., and was made to accommodate a design pressure of 1,850 lb./sq. in. The forged steel drum was constructed from two hollow forged half drums, each with an integral closed end.

NICKEL ALLOY STEELS

ensure reliability in lathes and gear shapers

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To ensure their products do not suffer through material failures, Messrs. Drummond Bros. Ltd., Guildford, specify nickel alloy steels in their 'Maxicut' industrial lathes and gear shapers and high production multi-tool machines.

Among the various nickel alloy steels used in Drummond tools is the 4¼ per cent nickel-chromium case-hardening steel, EN 39, which is used for worms and has adequate toughness and wear resistance for this exacting service.

TYPICAL CORE MECHANICAL PROPERTIES OF

EN39

SIZE	HEAT TREATMENT	MAXIMUM STRESS t.s.i.	ELONGATION per cent	IZOD ft. lb.
1¼" dia.	Oil quenched 830°C. tempered 160°C.	88·2	13	38
2" dia.	Oil quenched 830°C. Oil quenched 760°C. tempered 180°C.	82·1	17·5	61
4" dia.	Oil quenched 830°C. Oil quenched 760°C. tempered 180°C.	77·7	16·5	62

The benefits to be gained from the more highly alloyed case-hardening nickel steels, such as EN 33, EN 34, EN 36 and EN 39 include ease of heat-treatment, minimisation of processing distortion, and general reliability.

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BOOKS

The structure of metals: a modern conception
Published for the Institution of Metallurgists by Iliffe & Sons Ltd., May, 1959. 118 pages, with 50 illustrations in the text and 12 pages of art plates. 25*s*. net (by post 25*s*. 10*d*.).
EVERY YEAR since 1947 the Institution of Metallurgists has held a refresher course for its members, during which four leading authorities survey the present state of knowledge in relation to a particular aspect of metallurgy. Originally these papers were republished in book form solely for the benefit of Institution members, but the information they contained was considered far too valuable for only a limited circulation. Therefore, in conjunction with the present publishers, it was arranged that all papers read at refresher courses from 1956 onwards should be made available to a wider public.

The theme of the 1958 refresher course was a review of some of the modern theories of the structure of metals. In the first paper, Professor Raynor deals with developments which have taken place over the past 20-30 years in the electron theory of metals, and concisely outlines modern concepts of this highly complex subject. Dr. Catterall, in the paper which follows, describes the experimental techniques which have been developed to test the theory and to establish such things as the width of the valence band; the shape of the ' density of states ' curves; the shape of the Fermi surface; the numbers and signs of charge carriers in electrical conduction; and the change of the foregoing properties with alloying.

Professor A. G. Quarrell discusses the dislocation theory of plastic deformation, which has completely revolutionized thought about the structure and mechanical properties of metals. In the final paper Dr. Nutting shows that this theory, at first received with some scepticism, has now been thoroughly justified by experiment, the techniques of which he describes.

Notes for authors
The Institute of Physics, 47 Belgrave Square, London, S.W.1. 1959. 36 pages. 3*s*. 6*d*. (including postage).
THIS REVISED EDITION of the Institute's booklet is intended to assist authors in the preparation of technical papers and gives hints on the preparation of scripts and diagrams, on the layout of mathematics, proof correction, etc. In addition to a bibliography of reference books on technical writing, lists are given of abbreviations and symbols which seem to follow closely the recommendations of the British Standards Institution. Although the booklet is primarily intended for authors preparing papers for the Institute of Physics, the recommended practice would largely apply to publication in most other journals.

Research for industry, 1958
A report on work done by the industrial research associations in the Government scheme. Published for D.S.I.R. by H.M.S.O. 7*s*. 6*d*. (8*s*. by post).
THE INDUSTRIAL GRANTS COMMITTEE, in their statement of policy on research associations, envisage that in the next five years the total grant to organizations in the Government scheme will be a third higher than in 1954-59—but that the total grant-earning income will go up by nearly a half. While this increase implies a big expansion of existing research associations the Committee anticipate one or two new R.A.s. They point out that nearly one-fifth of net output of British manufacturing industry is not covered properly by existing research facilities and that some of these trades may in the future be appropriately served by grant-aided research bodies.

The publication *Research for Industry* made its debut last year, when the Council for Scientific and Industrial Research separated the section dealing with the R.A. movement from its own annual report. This year the new publication contains, in addition to the Industrial Grants Committee's statement of policy, a review of achievements, two special articles on research topics, and a complete catalogue of the R.A.s, describing the scope of each.

British Iron and Steel Federation, annual report 1958
Copies available from the Information Office, Steel House, Tothill Street, London, S.W.1.
THE ANNUAL REPORT of the British Iron and Steel Federation for 1958 makes the following interesting points:

Last year was a difficult year for the British steel industry, which felt the full effect of the general slackening of economic activity in the U.K. and most other industrialized countries. Although the capacity of the British steel industry continued to expand, crude steel production fell by over 2 million ingot tons from the 1957 record total to 19·6 million ingot tons. There was a progressive decline in the demand for steel from the home market and exports did not reach the high level attained in 1957. Nevertheless, the British steel industry retained its share of the world market in steel. There was significant saving in imports in 1958.

The set-back in demand is regarded as a purely temporary pause in the long term rising trend. The industry is continuing to plan its development accordingly. Capital expenditure by the industry was at a record rate of £105 million in 1958 compared with £95 million in 1957 and £75 million in 1956.

Press bending
By G. E. Hart. Chamberlain Industries Ltd., Staffa Works, Staffa Road, Leyton, London, E.10
A USEFUL little booklet providing information on all aspects of pipe bending for steam, gas and copper tubes. In particular it will serve as a manual of the ' Staffa ' two-stage hydraulic pipe-bending machine.

Die Fertigungsbelange des Gesenkschmiedens und ihre wissenschaftliche Weiterentwicklung
By Dr. Ing. Kurt Lange (Hannover), Springer Verlag, 1957.
THE BOOKLET deals with the technical and economic aspects of drop forging. After a brief discussion of the various theories the author points to the many problems which must be solved. This is a useful little book written in clear terms, although in German, and should prove useful to the English student of metallurgy.

U. E.

Aluminium rainwater goods
Aluminium Development Association, Application Brochure 13, 1959. 10*s*. 6*d*.
ALUMINIUM RAINWATER GOODS, both wrought and cast, are now established both by the existence of the relevant British Standard and by their approval and use by local authorities and their specification by architects. Illustrations show examples of installation on various types of building and include several case histories. The principal section of the book deals with dimensions of the main types of gutters, *i.e.* half-round, ogee and rectangular, also with the associated fittings and accessories. Rainwater pipes and fittings are similarly treated. Notes are given on the properties of aluminium base materials relevant to this application.

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Surface grinding machines

An entirely new range of surface grinding machines, identified as Wickman-Scrivener 618 and 824, marks yet another stage in the development of Scrivener precision grinding machines by Wickman Ltd., Coventry, in association with the parent company, Wickman Ltd. The machines are of radically new design, incorporating many entirely new features. They are built in two sizes, 6 in. by 18 in. and 8 in. by 24 in., generous proportioning ensures maximum support under all conditions of use. The attainment of exceptional precision is further assisted by basic machine elements of great stiffness and rigidity. The cross-feed handwheel graduations are in units of 0·001 in. (0·02 mm. on metric machines), and in addition a fine cross-feed control is calibrated in units of 0·0001 in. (0·002 mm.). Table speeds are high, permitting unusual stock removals, thus greatly reducing grinding times besides being invaluable for cool-grinding thin materials without distortion. The forged nitralloy spindle is of ample cross-section, giving great stiffness, rigidity and long life.

The whole of the hydraulic valve gear controlling the table and cross-feed movements is contained in a single monobloc assembly attached to the lower surface of the saddle, which is readily accessible and equally readily removable. The controls are conveniently grouped so that there is no need for the operator either to move or to stretch when operating the machine. Lubrication is entirely divorced from the hydraulic system, filtered lubricant being automatically metered to the table ways at all times, whether the table is being operated manually or under power.

Arc-welding electrodes

Six new types of manual arc-welding electrodes have recently been added to their comprehensive range by the English Electric Co. Ltd. 'Pyristees' have been developed for welding creep-resisting low-alloy steels, particularly where these are used for steam pipes operating at elevated temperatures; they cover working temperatures up to 1,050° F. (565° C.). 'Weldees' electrodes are all-position general-purpose electrodes for welding mild steel. They have the great advantage that they can be used with the 'touch' technique in all positions. 'Manganees' electrodes have been developed particularly for the welding of 12-14% manganese steel components. This is their main application although they can be used for reinforcing purposes. They are also suitable for welding 12-14% manganese steel to carbon steel and when used for surfacing carbon steel the resultant deposit has a hardness figure of 195-215 V.P.H., which work-hardens in service to approximately 500 V.P.H. 'Pressurees' I.P.1 are of the iron powder type of electrode specially developed to produce extremely fast welding with very good penetration properties in the downhand position. They are recommended for the welding of deep grooves and tight fillets, from which the inflated slag can easily be removed. '6011' electrodes, as the title suggests, have been designed for applications where an electrode having the American classification E.6011 would be used, particularly when welding oil pipelines on site. 'Herculees' 35 electrodes are of the low-hydrogen type and have been produced for welding heavy sections and highly-restrained joints in mild steel, medium high-tensile steels, carbon and alloy steels, and are also recommended for sulphur-bearing steels, producing welds free from porosity and cracking. The slag produced is easily removed and manipulation in all positions is very easy.

Dry film lubrication

Acheson Colloids Ltd. are now manufacturing and marketing a new range of resin-bonded dispersions of PTFE under the trade mark 'EMRALON.' These dispersions represent an advance in the technique of dry film lubrication and surface coating.

Polytetrafluoroethylene is already well known as a lubricating solid with the following outstanding properties: (1) Extremely low coefficient of friction (only 0·04 under certain conditions); (2) good stability at high temperatures; (3) high resistance to chemical attack; and (4) excellent 'parting' properties.

These advantages encouraged the use of thin films of pure PTFE as dry film lubricants and parting agents on metal and other surfaces. One major disadvantage of pure PTFE, however, is the need for a very high sintering temperature, for it must be raised to approximately 400° C. before it will coalesce and form a film capable of adhering to a surface. For this reason PTFE is not easily applied to plastics, rubber, wood or certain metals likely to be affected by the high sintering temperatures.

The important advantages of 'EMRALON' dispersions over pure PTFE films are: (1) They can be applied to heat sensitive surfaces; (2) a much improved adhesion is obtained; (3) the coatings have greater resistance to wear; and (4) they are more resistant to corrosion.

Slip casting

'Slip casting,' a process used for some 200 years in the ceramics industry, is being investigated anew as a possible answer to the problem of how to form brittle refractory metals for high-temperature applications. In this process, a powdered material is suspended in a liquid, and the resulting 'slip' is poured into an absorbent mould. The mould absorbs the liquid, and leaves a casting strong enough to be handled during drying and firing. In the case of metal suspensions, this deceptively simple process is complicated by the fact that the behaviour of the casting is affected by the composition and physical characteristics of the slip through a variety of rapidly occurring and little-understood chemical reactions. This is now being investigated in detail by the General Electric Research Laboratory, New York. In the past, slip casting has been performed with plaster of Paris moulds, but in the course of current studies it has been discovered that ordinary foundry sand moulds may also be used. As a result it is possible to construct intricate shapes easily, and a number of problems related to removing the finished part from the mould are solved.

Cadmium

Cadmium has become important for many specialized uses, including the atomic energy programme, according to a Bureau of Mines publication released by the U.S. Department of the Interior.

All commercial cadmium is a by-product of ores containing zinc. Accordingly, cadmium output is determined largely by zinc production instead of by price and demand. World production of cadmium is about 18 million lb. annually. The U.S. now produces approximately three-quarters of this, much of it through treating imported ores and concentrates.

The Bureau publication traces the development of uses for cadmium, by far the most important of which quantity-wise is electroplating, followed by pigments and low-melting-point alloys such as are used in automatic sprinkler systems. Other uses are in cadmium-nickel batteries and as control rods in nuclear reactors.

Electrical Aids in Industry *Data Sheet* **No. 6**

Lighting-1

Good lighting affects productivity in three ways, (1) by directly increasing the speed of working and reducing errors and wastage, (2) by improving overall factory efficiency through better supervision and housekeeping, (3) by providing better working conditions, thus improving labour relations and avoiding frequent changes of workpeople.

It is not possible to judge by the eye alone whether the lighting in any factory is good enough to serve these purposes because the eye has a tremendous range of adaptation (vision of some kind is possible within an illumination range of 1,000,000 to 1) and is hence an unreliable measuring instrument. Severe mental and eye strain or unconscious slackening of working speed may occur under lighting which *appears* to be adequate.

The only reliable way to appraise factory lighting is to conduct a lighting survey using a lightmeter an inexpensive instrument which measures the actual illumination available. These figures can then be compared with official recommended values for the various tasks.

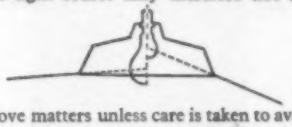
NATURE OF WORK, PROCESS OR MATERIAL	ILLUMINATION LUMENS/SQ.FT.
Rough or routine work. Large detail. Medium to light material of good contrast.	7
Ordinary work usually involving workers' inspection. Medium detail and contrast.	15
Skilled work, small detail or dark material.	30
Fine or critical work, very small detail, very poor contrast or very dark material.	50
Very fine exacting work.	100

In addition, readings should be taken at selected working points with the lightmeter so placed that it measures the light

on the work while the operative is working. This will indicate whether full use is being made of the light or whether shadow is obscuring part of it.

Glare

Any direct light on the eye tends to reduce its sensitivity — thus reducing the power to see. Increasing the intensity of the light source may therefore not improve matters unless care is taken to avoid glare either by correct positioning or by the use of correctly matched reflectors and lamps.



Shadows

Unnecessary shadows may seriously slow down work and also cause accidents. Lightmeter readings should always be taken under conditions exactly similar to those obtaining while work is going on. The operative himself may mask his work, an overhead crane or a heavily loaded conveyor belt may periodically obscure a light fitting, or a dust-laden atmosphere may reduce the designed lighting values.

For further information, get in touch with your Electricity Board or write direct to the Electrical Development Association, 2 Savoy Hill, London, W.C.2. Telephone: TEMple Bar 9434

Excellent reference books are available on electricity and productivity (8/6 each or 9/- post free) — "Lighting in Industry" is an example.

E.D.A. also have available on free loan a series of films on the industrial uses of electricity. Ask for a catalogue.

7659-4

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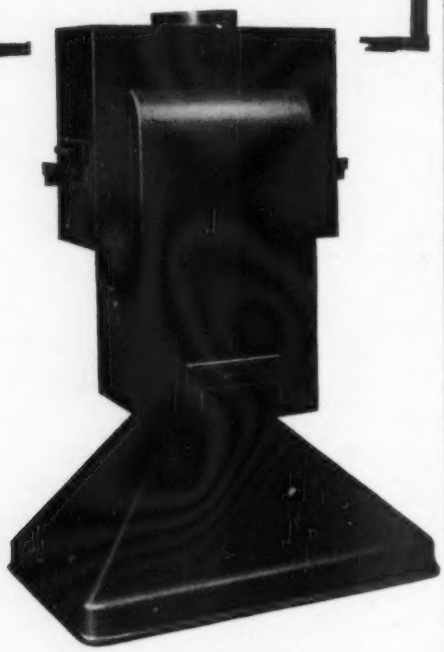

This photograph shows an Alldays & Onions UF/1 Dust Collector, primarily for use with hearths and designed to meet the requirements of the Clean Air Act.

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Vanadium	·02	·04
Molybdenum	*below* ·01	
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Cobalt	·005	·008
Antimony	*below* ·01	
Tungsten	*below* ·01	
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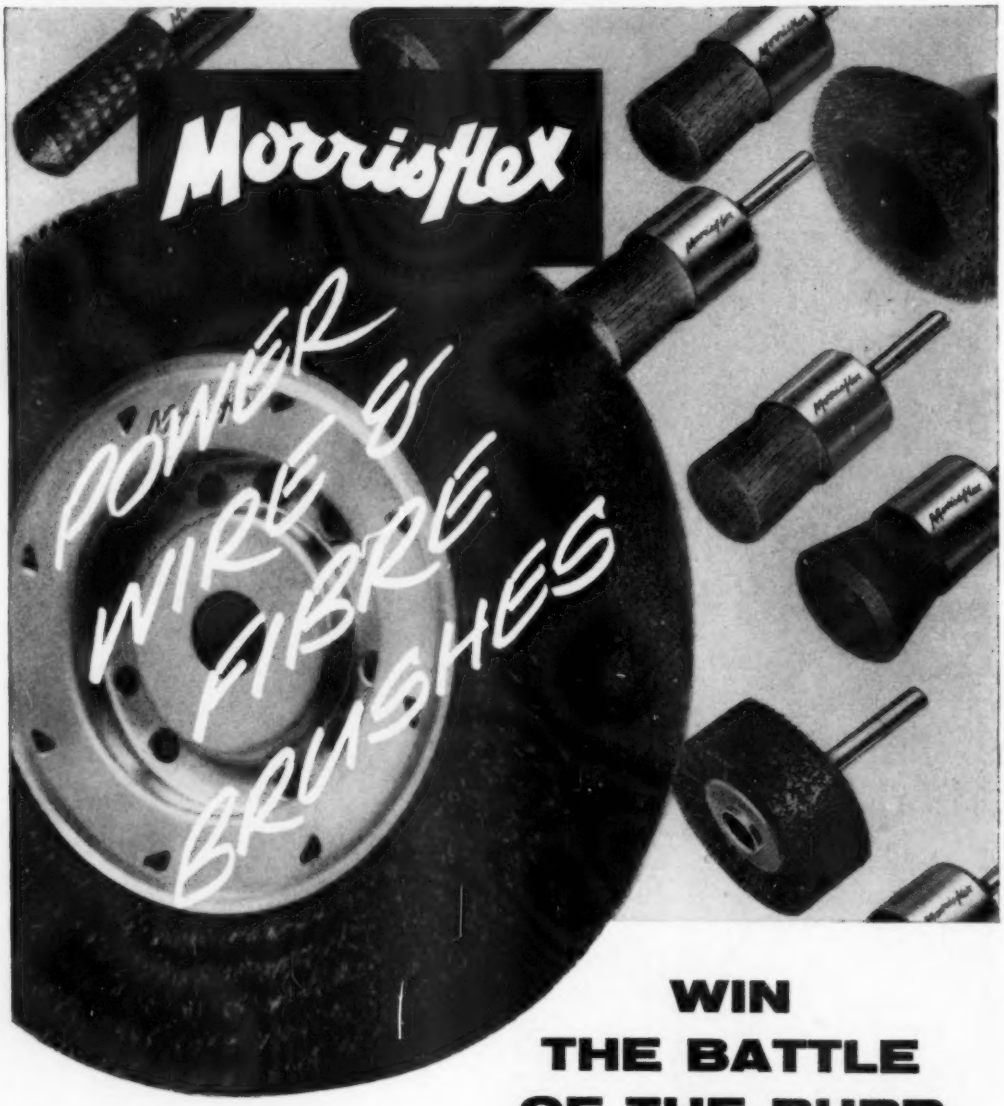

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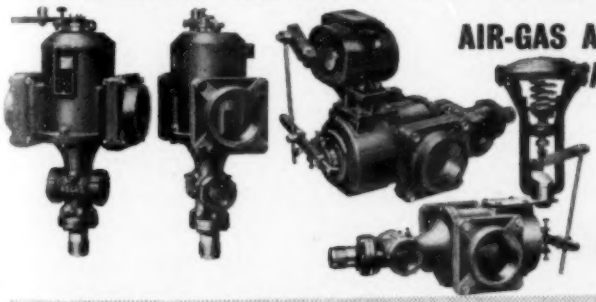

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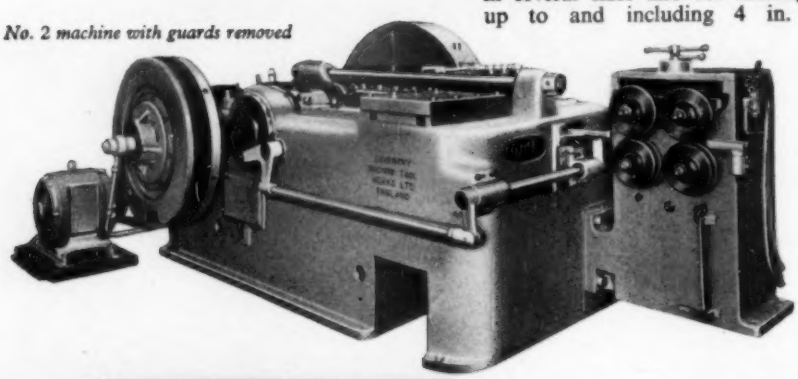
No. 2 machine with guards removed



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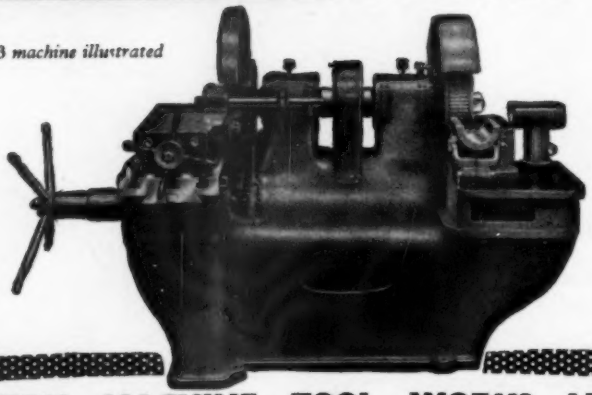
No. 3 machine illustrated



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Some technical facts.

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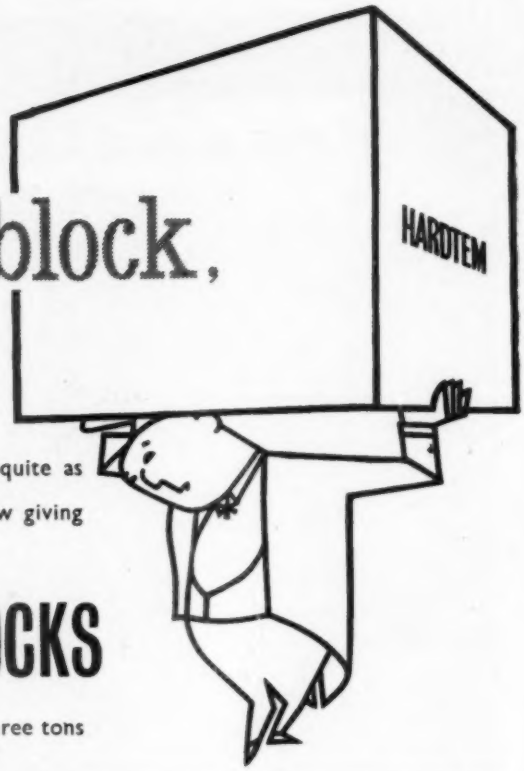

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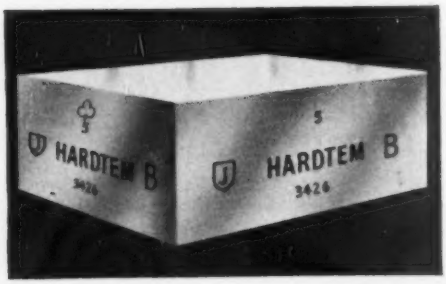

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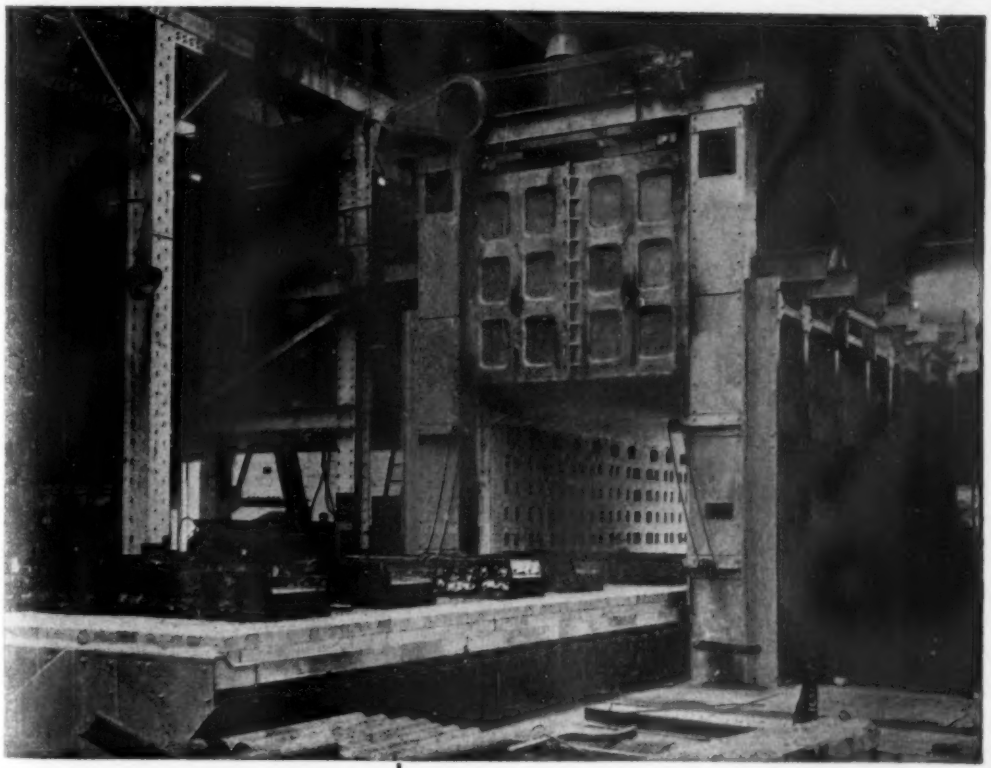

Photograph by courtesy of Messrs. T. Firth & J. Brown Ltd., Sheffield

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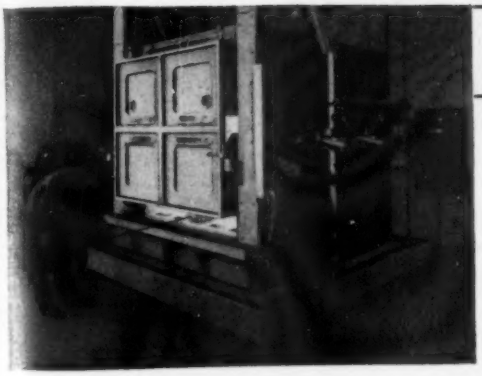

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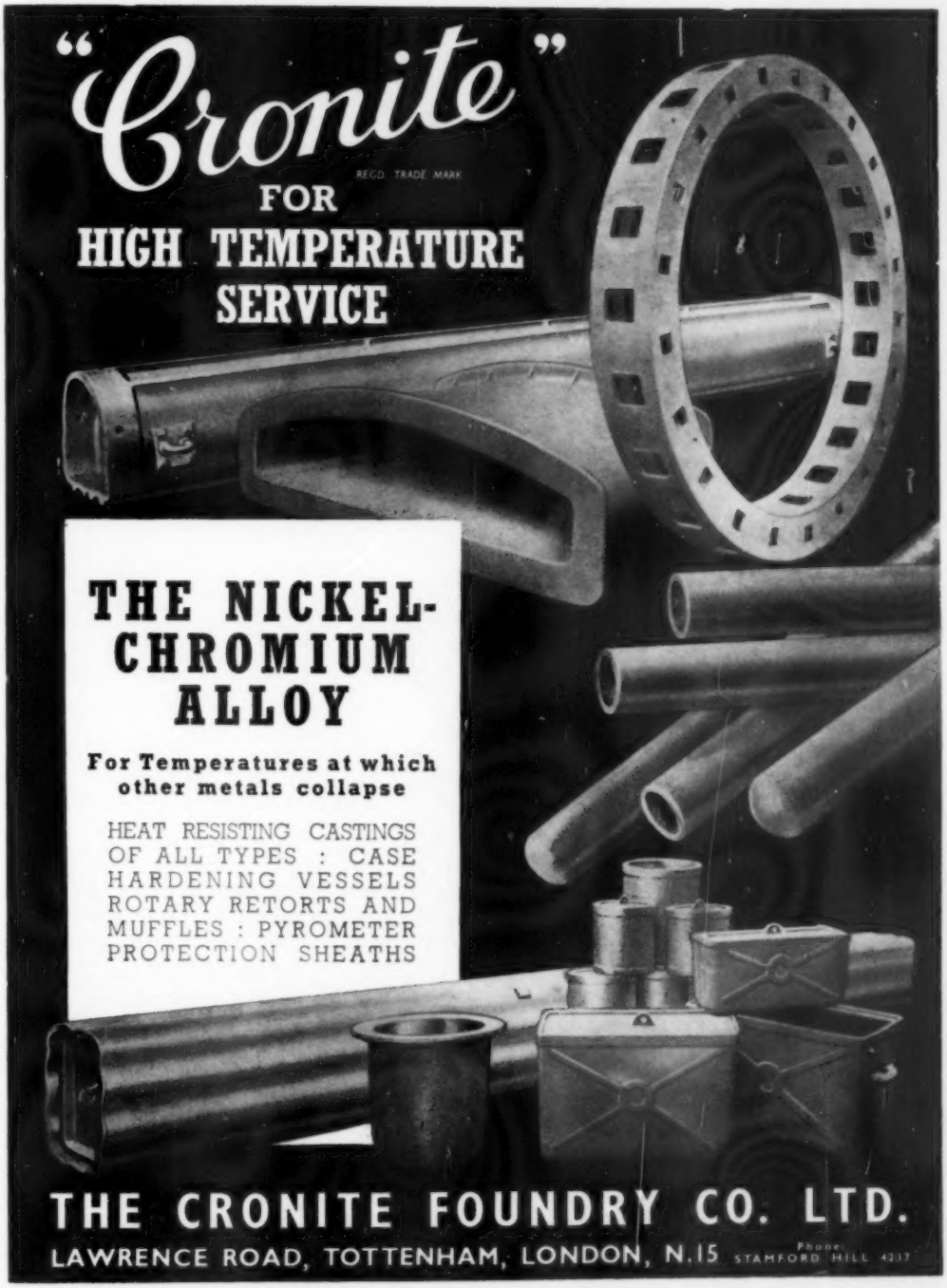
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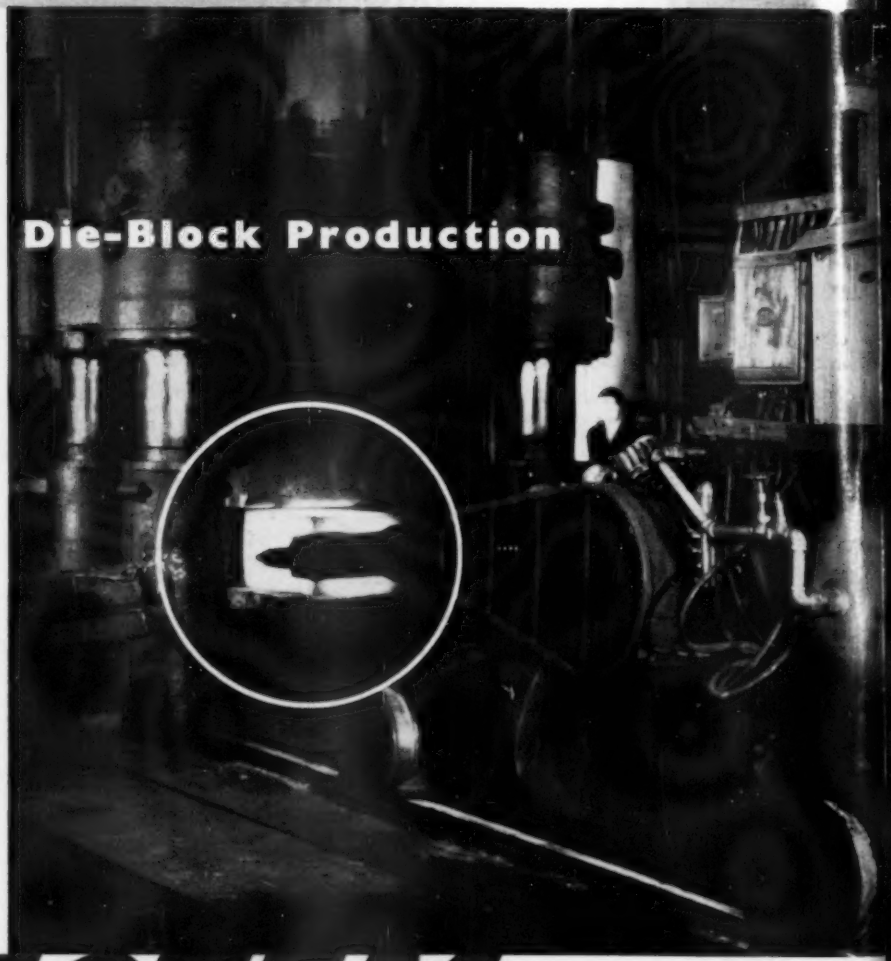

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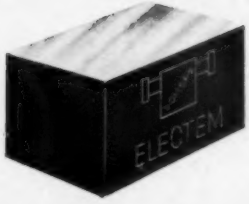

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